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PROCEEDINGS OF THE ANNUAL MECHANICS OF COMPOSITES REVIEW (15TH)

Sponsored by:

Air Force Wright Aeronautical Laboratories
Materials Laboratory

APRIL 1997

FINAL REPORT FOR PERIOD 24-25 OCTOBER 1990

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MATERIALS DIRECTORATE
WRIGHT LABORATORY
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13. ABSTRACT *(Maximum 200 words)*

This report contains the basic unedited vu-graphs of the presentations at the "Mechanics" of Composites Review" sponsored jointly by the Non-metallic Materials Division of the Air Force Materials Laboratory, the Structures Division of the Air Force Flight Dynamics Laboratory and the Directorate of Aerospace Sciences of the Air Force Office of Scientific Research. The presentations cover current in-house and contract programs under the sponsorship of these three organizations.

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epoxy-matrix composites; composite materials; resin matrix composites; composite bonded joints; fatigue of graphite/epoxy composites; fracture and fatigue of bi-materials

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FOREWORD

This report contains the abstracts and viewgraphs of the presentations at the Fifteenth Annual Mechanics of Composites Review sponsored by the Materials Laboratory. Each was prepared by its presenter and is published here unedited. In addition, a listing of both the in-house and contractual activities of each participating organization is included.

The Mechanics of Composites Review is designed to present programs covering activities throughout the United States Air Force, Navy, NASA, and Army. Programs not covered in the present review are candidates for presentation at future Mechanics of Composites Reviews. The presentations cover both in-house and contractual programs under the sponsorship of the participating organizations.

Since this is a review of on-going programs, much of the information in this report has not been published as yet and is subject to change; but timely dissemination of the rapidly expanding technology of advanced composites is deemed highly desirable. Works in the area of Mechanics of Composites have long been typified by disciplined approaches. It is hoped that such a high standard of rigor is reflected in the majority, if not all, of the presentations in this report.

Feedback and open critique of the presentations and the review itself are most welcome as suggestions and recommendations from all participants will be considered in the planning of future reviews.

Deborah Perdue

DEBORAH PERDUE, Meeting Manager
Mechanics & Surface Interactions Branch
Nonmetallic Materials Division
Materials Laboratory

ACKNOWLEDGEMENT

We wish to express our appreciation to the authors for their contributions; to the focal points within the organizations for their efforts in supplying the program listings; and to Barbara Woolsey for managing registration.

ADVANCES IN COMPUTATIONAL SIMULATION OF COMPOSITES BEHAVIOR

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216/433-3252

ABSTRACT

Predicting composite behavior is generally very complex because of the several inherent scales in its physical make-up. For example,these scales include: micromechanics (intraply), macromechanics (interply), laminate (several plies), local region (plate type finite element) and structural component (assemblage or many finite elements). There are two general methods for predicting the composite behavior in all its inherent scales: (1) continuum mechanics - the classical approach with heavy reliance on applied mathematics, and (2) the discretized methods such as finite difference and finite element. Briefly, in these approaches the specific behavior is formulated by using the participating variables that describe the physics and fundamental mechanics concepts to derive the governing field equations. Subsequently, the field equations are manipulated by using formal mathematical methods to reduce the number of unknowns to a manageable number that can be readily solved by available solution algorithms with the aid of the computer.This approach can justifiably be called "Computer Aided Solution."

An alternate approach is to simultaneously solve the fundamental governing field equations for all the participating variable by using the computer as integral part of the solution. Since this approach can be used to simulate behavior or process as well as a specific event it is called "Computational Simulation." Computational simulation has been successfully applied at Lewis Research Center to predict the behavior of polymer composites (ICAN, ref. 1) and composite structures (COBSTRAN, STAEBL, STAEBL/GENCOM, refs. 2,3), and more recently the behavior of metal matrix composites (METCAN,HITCAN, ref. 4).

The objective of this review is to summarize recent progress and present typical but important results. The specific topics covered in the review include: (1) high temperature composites behavior, (2) composites "structures" progressive fracture, (3) composite damping, and (4) probabilistic composite response. References to relevant reports are provided where more detailed information is available. Collectively, the results presented demonstrate that computational simulation is an effective approach to predict composite behavior in all its inherent scales, including: high temperature composite behavior, progressive fracture, tailoring fabrication processes and probabilistic composite structural analysis.

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1.0 Murthy, P. L. N. and Chamis, C. C.: Integrated Composite Analyzer. NASA TM 83700, May 1984.

2.0 Aiello, R. A. and Chamis, C. C.: Composite Blade Structural Analyzer (COBSTRAN) Theoretical/Programmer's Manual. NASA TM 101958, August 1989.

3.0 Rubenstein, R. and Chamis, C. C.: STAEBL/General Composites with Hygrothermal Effects (STAEBL/GENCOM). NASA TM 100266, December 1987.

4.0 Singhal, S. N.; Lackney, J. J.; Chamis, C. C.; and Murthy, P. L. N.: Demonstration of Capabilities of High Temperature Composites Analyzer Code HITCAN. NASA TM 102560, March 1990.

RELEVANT REPORTS

Chamis, C. C.; Murthy, P. L. N.; and Hopkins, D. A.: Computational Simulation of High Temperature Metal Matrix Composites Cyclic Behavior. NASA TM 102115, November 1988.

Caruso, J. J.; Chamis, C. C.; and Brown, H. C.: Parametric Studies to Determine the Effect of Compliant Layers on Metal Matrix Composite Systems. NASA TM 120465, July 1989.

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Chamis, C. C. and Murthy, P. L. N.: Computational Simulation of Structural Fracture in Fiber Composites. NASA TM 102505, December 1989.

Minnetyan, L.; Chamis, C. C.; and Murthy, P. L. N.: Structural Behavior of Composites with Progressive Fracture. NASA TM 10237, 1990.

Minnetyan, L.; Murthy, P. L. N.; and Chamis, C. C.: Progression of Damage and Fracture in Composites Under Dynamic Loading. NASA TM 103118, April 1990.

Saravanos, L. and Chamis, C. C.: Tailoring of Composite Links for Optimal Damped Elasto-Dynamic Performance. NASA TM 102094, September 1989.

Saravanos, D. A. and Chamis, C. C.: Mechanics of Damping for Fiber Composite Laminates Including Hygro-Thermal Effects. NASA TM 102329, April 1989.

Saravanos, D. A. and Chamis, C. C.: An Integrated Methodology for Optimizing Structural Composite Damping. NASA TM 102343, December 1989.

Thanedar, P. B. and Chamis, C. C.: Composite Laminate Tailoring with Probabilistic Constraints and Loads. NASA TM 102515, January 1990.

Chamis, C. C. and Stock, T. A.: Probabilistic Simulation of Uncertainties in Composite Uniaxial Strengths. NASA TM 102483, February 1990.

ADVANCES IN COMPUTATIONAL SIMULATION OF COMPOSITES BEHAVIOR

by
Christos C. Chamis
NASA Lewis Research Center
Cleveland, OH 44135

Fifteenth Annual Mechanics of Composites Review
October 24-25, 1990
Dayton, OH

BACKGROUND:

COMPOSITES BEHAVIOR IS GENERALLY COMPLEX BECAUSE OF THE MULTI-INHERENT SCALES AND BEHAVIOR COUPLING IN INTRA- AND INTER-SCALES. THAT FOR HIGH TEMPERATURE METAL-MATRIX COMPOSITES IS COMPOUNDED BECAUSE OF THE PRESENCE OF NONLINEAR DEPENDENCIES:

- o TEMPERATURE
- o STRESS
- o TIME

VARIOUS APPROACHES CAN BE USED TO DESCRIBE COMPOSITE BEHAVIOR WITH VARIOUS DEGREES OF SOPHISTICATION:

- o CONTINUUM MECHANICS
- o DISCRETE METHODS
- o COMPUTATIONAL SIMULATION

BACKGROUND (CONTINUED):

COMPUTATIONAL SIMULATION HAS BEEN PURSUED AT LEWIS OVER THE PAST TWO DECADES BECAUSE:

- o FORMULATIONS ARE SIMPLE
- o INCLUSIVE OF PHYSICS
- o NONLINEARITIES EASILY INCORPORATED
- o COMPUTER PERFORMS ALL REQUIRED COMPUTATIONS FOR THE SIMULATION

OBJECTIVE:

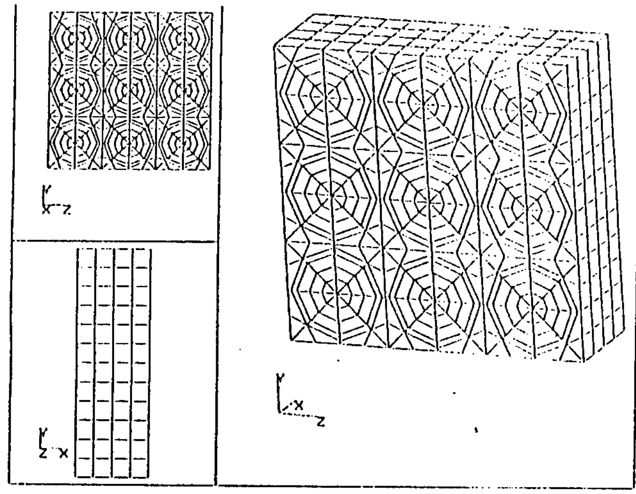
SUMMARIZE RESEARCH ACTIVITIES/PROGRESS ON COMPUTATIONAL SIMULATION OF COMPOSITES BEHAVIOR AND PRESENT TYPICAL RESULTS.

COMPUTATIONAL SIMULATION

- o HIGH TEMPERATURE COMPOSITES BEHAVIOR
- o COMPOSITE "STRUCTURES" PROGRESSIVE FRACTURE
- o COMPOSITE DAMPING
- o PROBABILISTIC COMPOSITE RESPONSE

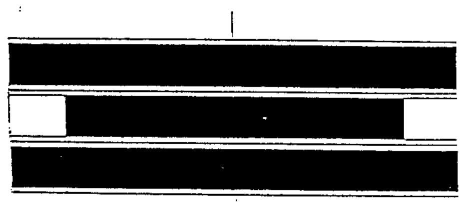
HIGH TEMPERATURE COMPOSITES BEHAVIOR

- INTERPHASE EFFECTS
- MICROFRACTURE
- TAILORED PROCESSING
- COMPUTER CODES
  - METCAN - METAL MATRIX COMPOSITE ANALYZER
  - HITCAN - HIGH TEMPERATURE COMPOSITE-STRUCTURAL ANALYZER
  - CEMCAN - CERAMIC MATRIX COMPOSITE ANALYZER



FINITE ELEMENT MODEL FOR METCAN VALIDATION AND CONSTITUENT MICROSTRESSES

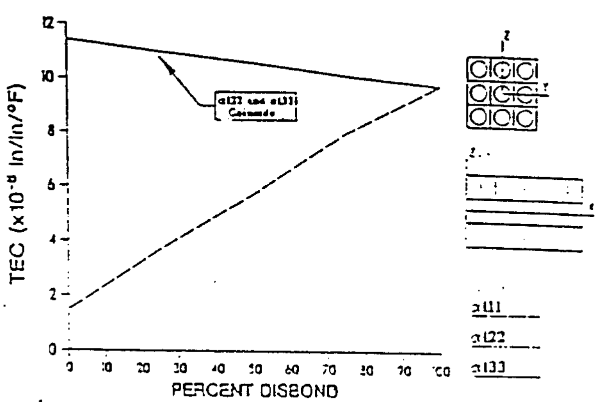
FIBER DISBONDING AND INTERPHASE (COMPLIANT LAYR) EFFECTS



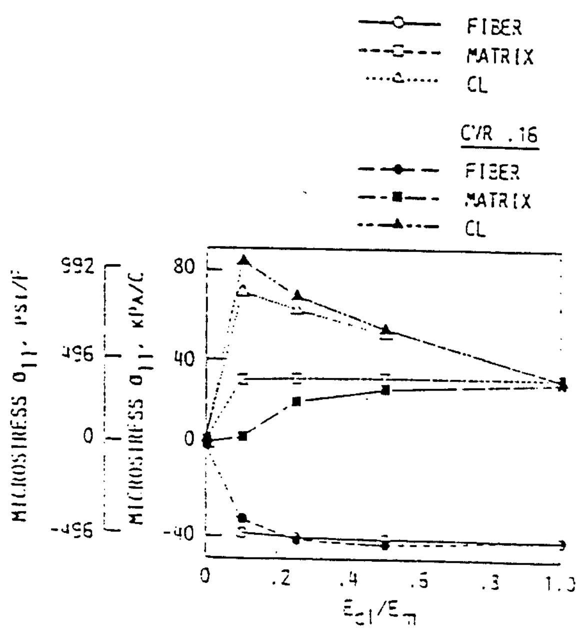
Plane of symmetry

Center fiber disbonding showing 2.73% disbonding

EFFECT OF FIBER DEBONDING ON THERMAL EXPANSION COEFFICIENT (TEC) – α111, α122, α133



Effects of nine fiber disbonding at room temperature on thermal expansion coefficients



ROOM TEMPERATURE CONSTITUENT MICROSTRESS $\sigma_{\ell 11}$ WITH VARYING CL CTE FOR 46 PERCENT FVR SCS6/Ti-15-3 AT L/D = 2.94.

TAILORED PROCESSING - HISTORY AND MINIMIZED MICROSTRESSES

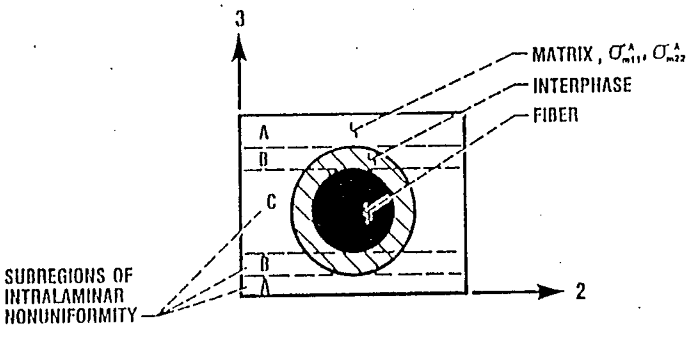
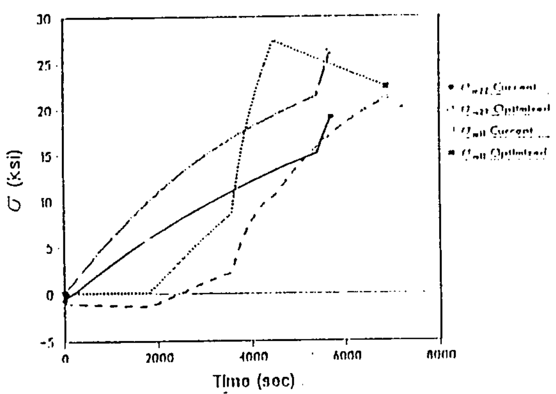
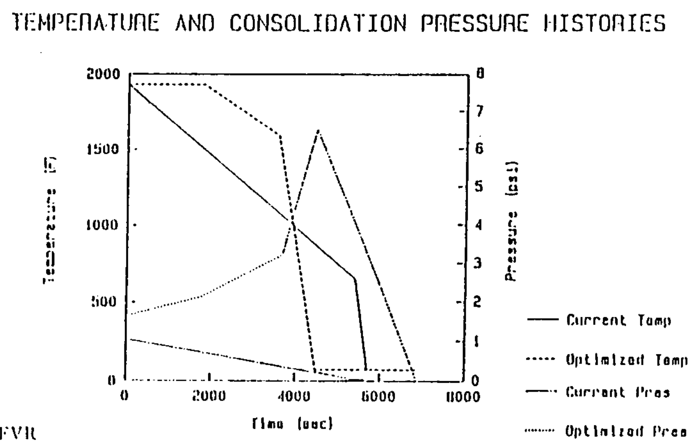


Fig. 1 Material microregions in a representative metal-matrix composite cell.

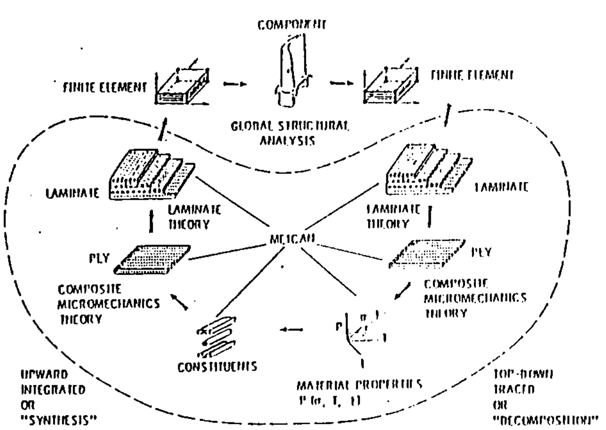


Matrix microstress developed during the cool-down phase of 0.40 FVR P100/Copper (Case 1). Optimum and current processes.

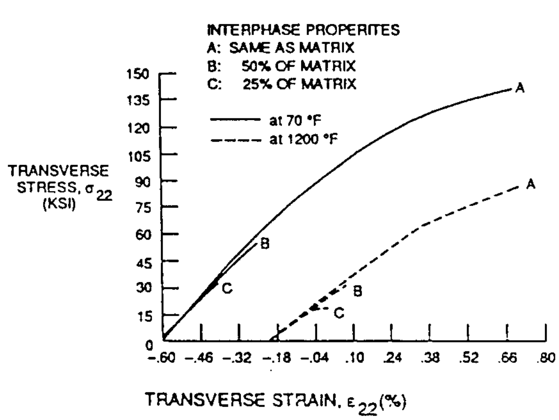


Optimum and current cool-down phases for 0.40 FVR P100/Copper (Case 1).

INTERPHASE EFFECTS ON PLY STRENGTH

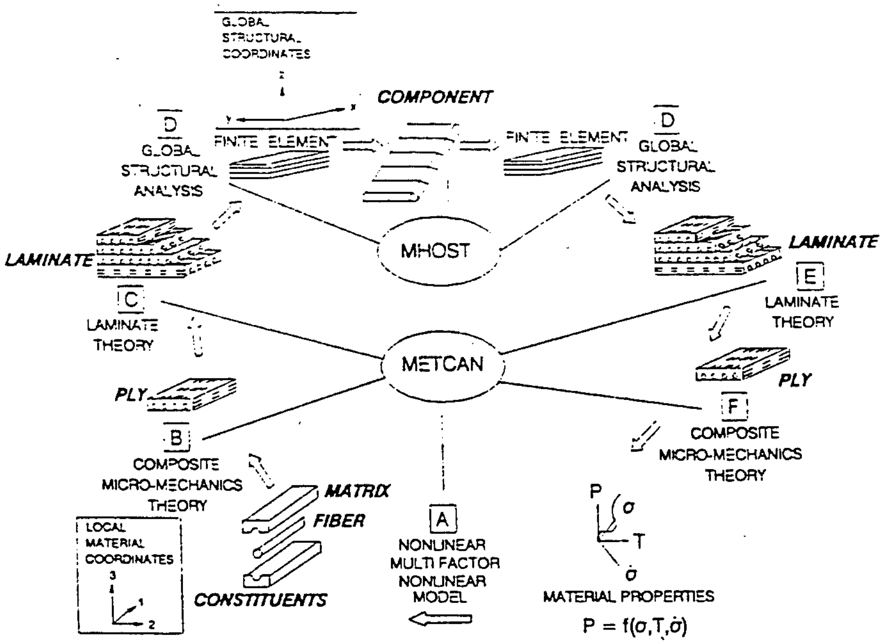


HIGH TEMPERATURE COMPOSITE BEHAVIOR COMPUTATIONALLY SIMULATED.



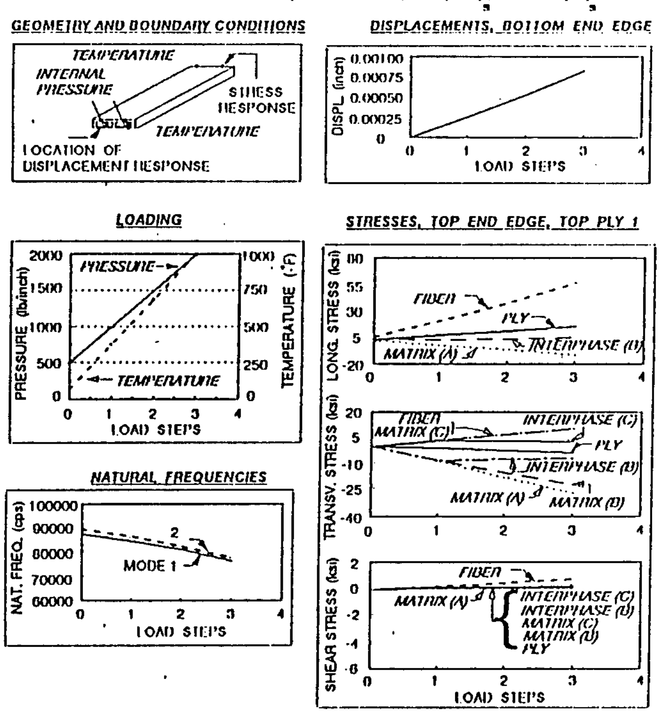
- Influence of interphase property degradation on the response of SiC/Ti-15-3-3-3 unidirectional composite in transverse tension.

HIGH TEMPERATURE COMPOSITE STRUCTURAL ANALISYS - BUILD-UP STRUCTURE



HITCAN: An Integrated Approach for High Temperature Composite Structural Analysis

BOTTOM SUPPORTED BUILT-UP STRUCTURE UNDER BENDING AND UNIFORM TEMPERATURE LOADINGS
FOR (SI C/TI-15-3-3-3, TOP:[90,0]$_s$, BOTTOM:[90]$_s$, SPARS:4[0]$_s$); 0.4 FIBER VOLUME RATIO



SENSITIVITY ANALYSIS AT LOAD STEP 3

EFFECTS OF	DISP. (BOTTOM END EDGE) (inch)	STRESSES, PLY 1 (TOP END EDGE) (ksi) LONG.	TRANS.	SHEAR
BASE CASE *	0.000823	15.2	3.3	0.1
FIBER DEGRADATION	0.000888	14.4	-2.0	0.1
FABRICATION	0.000420	12.2	0.5	0.02
NONLINEAR MULTIFACTOR INTERACTIVE CONSTITUTIVE MODELS				
P = CONSTANT	0.000710	15.2	2.0	0.1
P = f(T)	0.000828	10.4	-3.0	1.0
P = f(σ)	0.000707	14.3	2.4	0.1
P = f($\dot{\sigma}$)	0.000710	15.2	2.0	0.1

* BASE CASE IS FOR NO FIBER DEGRADATION, NO FABRICATION, AND P = f(T, σ, $\dot{\sigma}$) CONSTITUTIVE MODEL.

NOTATION:

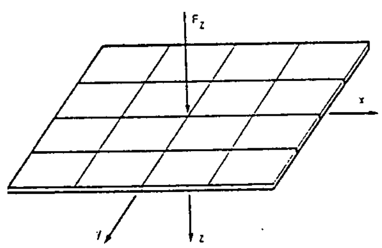
P = MATERIAL PROPERTY *T = TEMPERATURE*
σ = STRESS *$\dot{\sigma}$ = STRESS*

- BASE CASE AND SENSITIVITY ANALYSIS RESULTS FOR BUILT-UP STRUCTURE

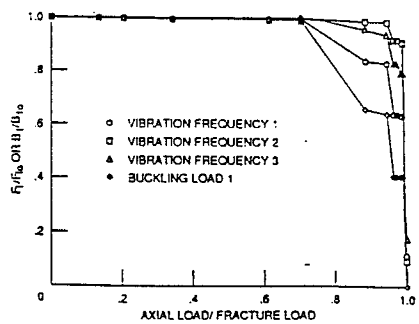
COMPOSITE "STRUCTURES" PROGRESSIVE FRACTURE

- STATIC LOADING
- DYNAMIC LOADING
- ENVIRONMENTAL EFFECTS

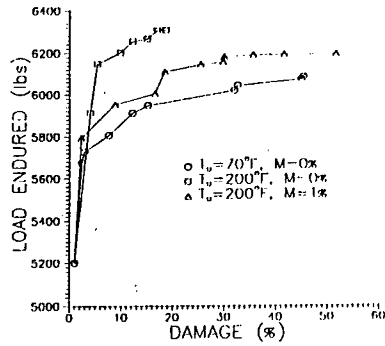
MONOTONIC STATIC LOAD: DRY AND HOT-WET ENVIRONMENT



- RECTANGULAR PLATE MODEL WITH NORMAL LOAD.

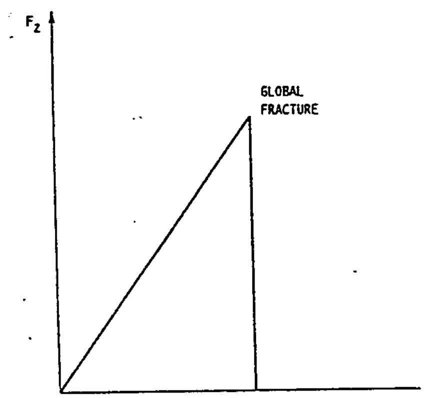


-Structural degradation for composite plate; support condition 1.

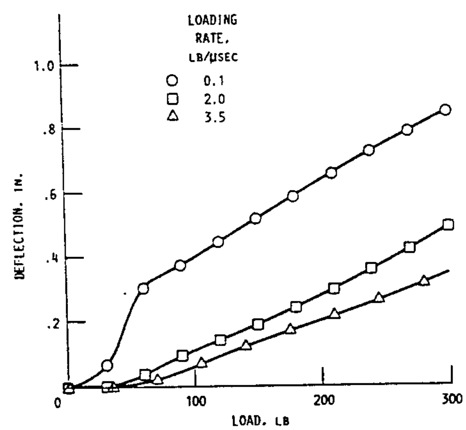


Created Damage with Applied Loading

MONOTONIC DYNAMIC LOAD

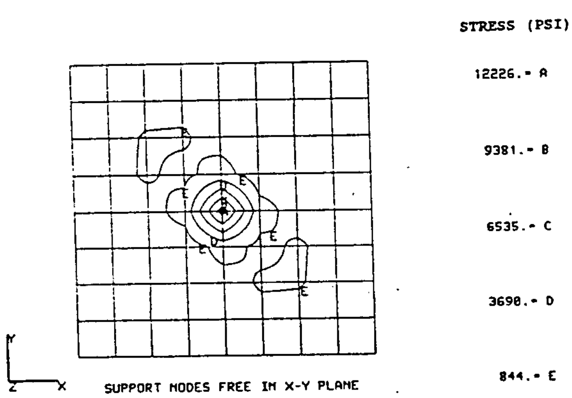


VARIATION OF APPLIED LOADING WITH TIME.



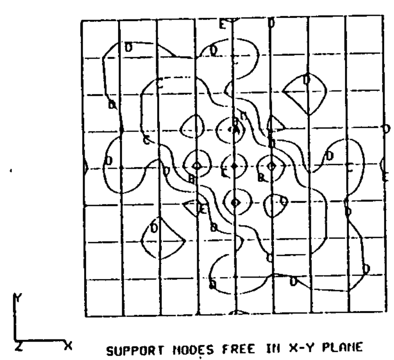
LOAD-DEFLECTION RELATIONSHIPS FOR RELEASED SUPPORTS.

LOADING RATE = 3.5 LBS/MICROSEC. APPLIED LOAD = 105 LBS. (B.E.)



TRANSVERSE STRESSES FOR PLY 1

TRANSVERSE STRESSES AT INITIAL DAMAGE.



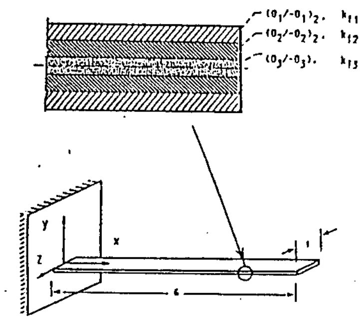
TRANSVERSE STRESSES FOR PLY 1

TRANSVERSE STRESSES AT DYNAMIC EQUILIBRIUM AFTER INITIAL DAMAGE.

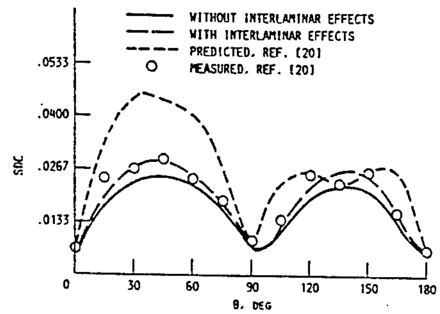
COMPOSITE DAMPING

- o ELASTO-DYNAMIC PERFORMANCE
- o LAMINATE DAMPING WITH HYGROTHERMAL EFFECTS
- o TAILORED STRUCTURAL COMPOSITE DAMPING
- o MULTI-OBJECTIVE STRUCTURAL OPTIMIZATION WITH DAMPING

SIMULATED LAMINATE DAMPING

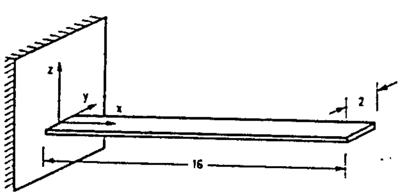


CANDIDATE COMPOSITE STRUCTURE. GRAPHITE/EPOXY (0.60 FVR, 70° F).

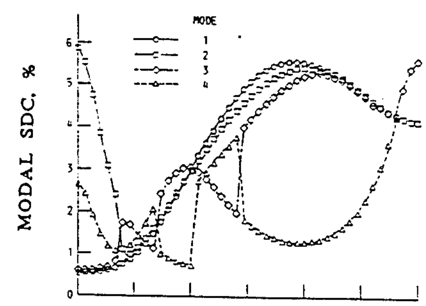


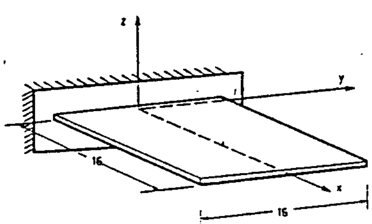
- FLEXURAL SDC OF A 50 PERCENT FVR HM-S/DX210 $(\theta/90+\theta/45+\theta/-45+\theta)_s$ LAMINATE.

SIMULATED STRUCTURAL DAMPING

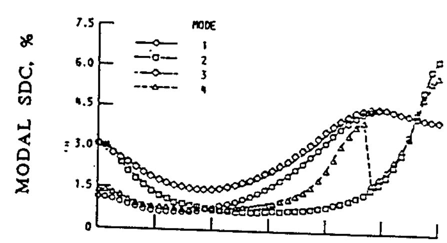


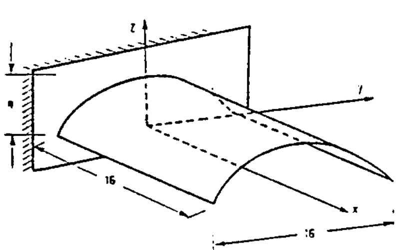
(a) CANTILEVER BEAM.





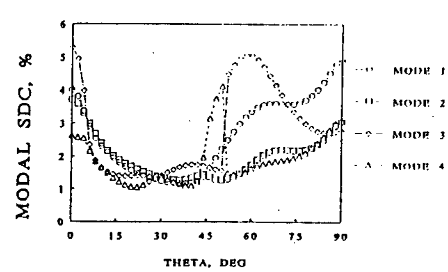
(b) CANTILEVER PLATE.





(c) CANTILEVER SHELL.

- THE THREE COMPOSITES STRUCTURES.

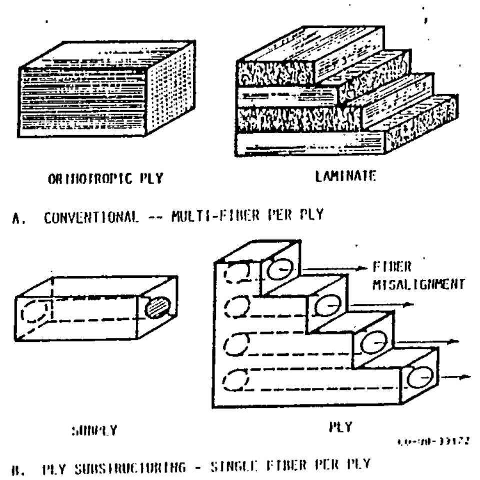


PROBABILISTIC COMPOSITE RESPONSE

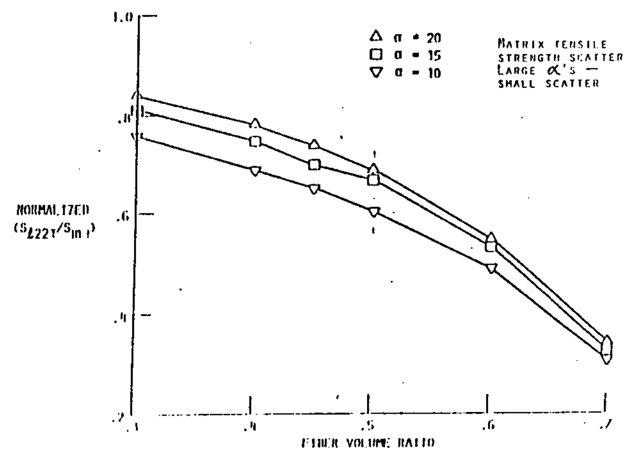
- o COMPOSITE STRENGTHS
- o LAMINATE TAILORING
- o STRUCTURAL ANALYSIS

PROBABILISTIC COMPOSITE STRENGTHS

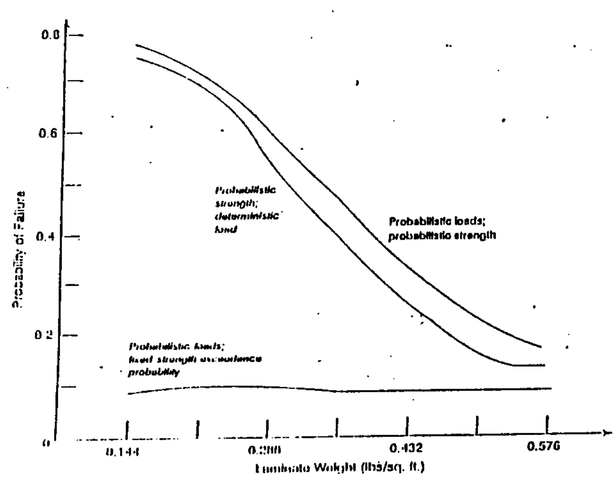
PLY SUBSTRUCTURING ANALOGOUS TO LAMINATE DECOMPOSITION FOR PROBABILISTIC COMPOSITE MECHANICS



MATRIX TENSILE STRENGTH SCATTER EFFECTS ON PLY TRANSVERSE TENSILE STRENGTH



PROBABILISTIC LAMINATE TAILORING



PROBABILISTIC BUCKLING ANALYSIS/RESPONSE

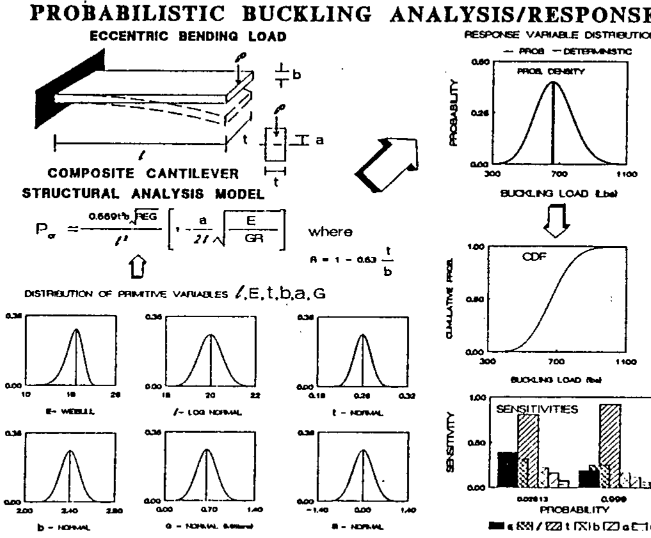


PLATE 5

CONCLUSIONS:

COMPUTATIONAL SIMULATION OF COMPOSITES BEHAVIOR HAS BEEN SUCCESSFUL FOR:

- o HIGH TEMPERATURE COMPOSITES
- o PROGRESSIVE COMPOSITE "STRUCTURES" FRACTURE
- o COMPOSITES DAMPING
- o PROBABILISTIC COMPOSITES RESPONSE

SPECIFIC FINDINGS:

HIGH TEMPERATURE COMPOSITES: COMPLIANT INTERPHASE SIGNIFICANTLY EFFECTS ON TRANSVERSE COMPOSITE PROPERTIES.

- o DEGREE OF INTERFACIAL BOND (EXCEPT NO BOND) - NOT SIGNIFICANT IN STRESS TRANSFER.
- o LONGITUDINAL THERMAL EXPANSION COEFFICIENT MOST SENSITIVE PROPERTY FOR INTERFACIAL HEALTH.
- o FABRICATION PROCESS CAN BE TAILORED USING SUITABLE MICROMECHANICS.
- o COMPUTER CODES:
  - METCAN - CONTINUING VERIFICATION
  - HITCAN - DEMONSTRATED FOR VARIOUS STRUCTURES/ FEATURES
  - CEMCAN - TWO LEVEL SUBSTRUCTURING TO CAPTURE LOCAL DETAILS

SPECIFIC FINDINGS (CONTINUED):

COMPOSITE "STRUCTURES" PROGRESSIVE FRACTURE:

- o STATIC LOAD - PROGRESSIVE FRACTURE DOES NOT INFLUENCE STRUCTURAL BEHAVIOR.
- o DYNAMIC LOAD - THE FASTER THE LOAD RATE, THE MORE LOCALIZED THE DAMAGE AND THE SMALLER THE GLOBAL DEFLECTIONS.
- o HYGROTHERMAL ENVIRONMENTS - INCREASE PROGRESSIVE FRACTURE FOR THE SAME LOAD CONDITIONS.
- o COMPOSITE DAMPING - CAN BE SIMULATED FROM MICRO-MECHANICS UPWARD AND CAN BE TAILORED FOR SPECIFIC APPLICATIONS.

SPECIFIC FINDINGS (CONTINUED):

- o PROBABILISTIC COMPOSITE RESPONSE:
  - PLY STRENGTH -- VARIOUS UNCERTAINTY EFFECTS CAN BE QUANTIFIED.
  - LAMINATE TAILORING -- CAN BE PERFORMED WITH PROBABILISTIC PROPERTIES, LOADS AND CONSTRAINTS.
  - STRUCTURAL ANALYSIS -- CAN BE EVALUATED PROBABILISTICALLY FOR ANY OR ALL RESPONSES.

E-5727

MICROFRACTURE IN HIGH TEMPERATURE METAL MATRIX COMPOSITES

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Microfracture (fiber/matrix fracture or interface debonding) in metal-matrix composites is critical to assess the fatigue resistance, durability, impact resistance and other important properties. In the present work, microfracture propagation and the extent of stress redistribution in the surrounding fiber/matrix due to fracture of fiber/matrix or interface debonding in a ply, has been simulated. This continuing research activity at NASA Lewis Research Center (Ref. 1) has led to the development of a computational simulation procedure, based on three-dimensional finite element analysis and strain energy release rates, to predict the microfracture process and identify/quantify the hierarchy of respective fracture modes in metal matrix composites.

The finite element model used in the computational simulation procedure consists of a group of nine fibers, all unidirectional, in three-by-three unit cell array ("nine cell model"). The procedure is illustrated by using a 0.35 fiber volume ratio SiC/Ti15 metal matrix composite. The load and boundary conditions are applied to the model through enforced displacements. Fracture is simulated by placing duplicate node points on either side of the crack. These duplicate nodal or grid points have the same geometrical location, but no connectivity exists between them, thus, in effect producing a defect of zero width. For an assumed fracture mode, uniform displacement boundary conditions are applied to the model in a given direction. Resulting nodal forces corresponding to those applied displacements are found by the finite element analysis. In a typical set of simulations, microfracture is initiated in a middle ply fiber and is allowed to grow either in the matrix or along the fiber-matrix interface. Eventually, it is allowed to grow in the interply regions and through the adjacent plies. Similarly, microfracture could be initiated in the matrix or in the interface. Comparison of resulting nodal forces is made for reduction in global stiffness. The corresponding strain energy release rates are computed for perturbed fracture configurations.

Strain energy release rate (SERR) is a commonly used indicator of the fracture toughness of a material. It gives a measure of the amount of energy required to propagate a defect in the laminate. Hence, one can make a direct comparison of the damage tolerances between different microfracture configurations (modes/paths), materials and geometries. One of the methods used to calculate strain energy release rate is the crack closure method. In this method, nodal displacements and the corresponding nodal forces at the crack tip location, are used to determine the amount of work required to close the crack, which has been increased by an incremental amount. This is a local level or microfracture approach since the amount of energy produced by the local displacements and forces at the crack tip, are used to calculate the corresponding strain energy release rate. However, in the present research, a global approach has been used to calculate the strain energy release rate. In this approach, applied nodal displacements and the corresponding nodal forces obtained from finite element analysis are used to calculate the work done. Strain energy release rate (G) is then, calculated as:

$$G = \frac{dW}{dA} = \frac{1}{2} \cdot \frac{F_2 \cdot u - F_1 \cdot u}{\Delta A}$$

where

- u applied displacement at the loaded end of the model
- dW change in work
- ΔA area of the new surfaces generated
- F_1, F_2 forces at the end nodes before and after ΔA

This equation is simply incremental change in work divided by the incremental change in new surface area that opens up. The applied displacements remain constant, but the resulting force required to maintain that displacement changes because of the reduction in stiffness of the composite as the fracture propagates.

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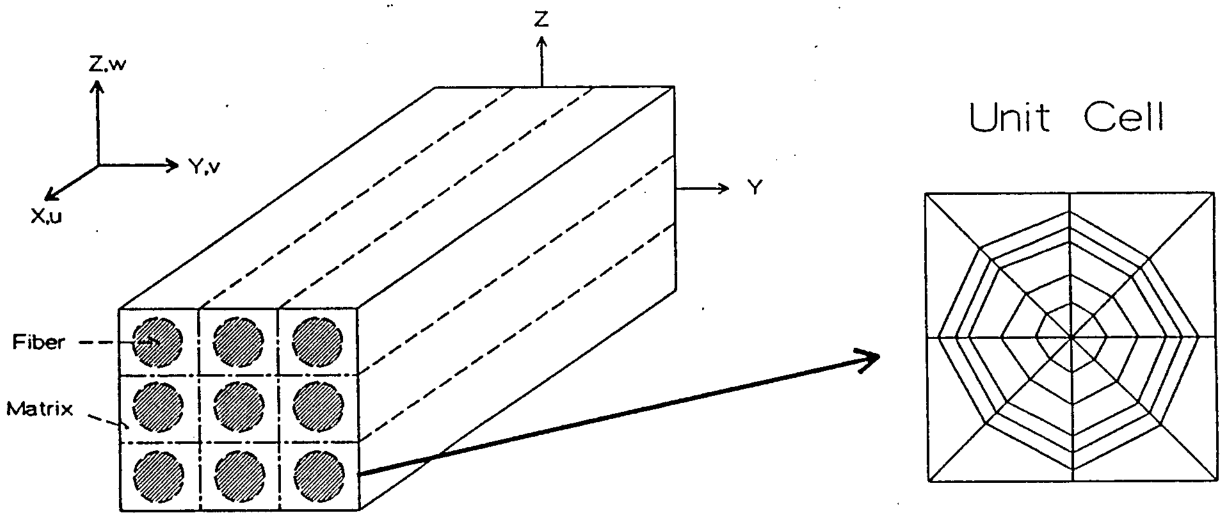
Typical results indicate that if the composite is subjected to longitudinal (along the fiber) loading, interface debonding does not initiate by itself. It instantaneously follows fiber (or matrix) fracture. Even though, it takes less energy to drive the crack in this fracture mode (debonding along fiber-matrix interface) under longitudinal loading, one could not reach this state prior to fiber (or matrix) fracture. The reduction in global longitudinal stiffness and corresponding strain energy release rates are negligibly small due to matrix fracture as compared to the fiber fracture case. It is also observed that under longitudinal loading, due to fracture in one fiber, the increase in longitudinal stress in neighboring fibers is small, implying that the random fracture in one fiber is unlikely to initiate fracture in neighboring fibers. Even if a substantial percentage of fibers are fractured in one plane causing a reduction in longitudinal strength in that plane, the reduction in global longitudinal stiffness is small and perhaps difficult to detect experimentally, at least for the composite system and fiber volume ratio investigated. Similarly, if the composite is subjected to transverse or shear type loading, debonding along the fiber-matrix interface is the only likely event for the fracture propagation. If debonding is present to begin with, transverse or shear loading will cause the extension in debonding. However, under shear loading, the composite is not as sensitive to debonding extension as it is under transverse loading. It is also observed that when the composite is subjected to bending loads, interply delamination is the only likely mode of fracture propagation under such loading. The model used and typical results are shown in the accompanying figures.

REFERENCES

1. Mital, S.K., Caruso, J.J., and Chamis, C.C.: Metal Matrix Composites Microfracture: Computational Simulation, NASA TM-103153, 1990

SCHEMATIC OF THE MODEL USED TO SIMULATE COMPOSITE MICROFRACTURE



PROPERTIES OF CONSTITUENT MATERIALS OF SiC/Ti15

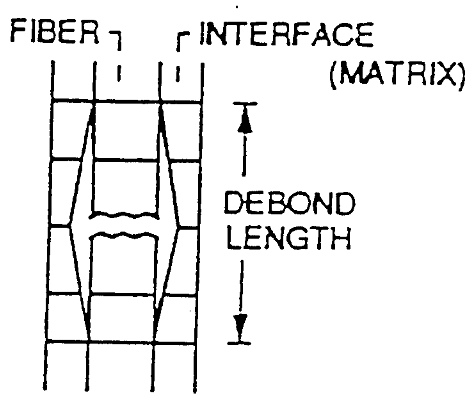
	SiC Fiber	Ti15 Matrix	Interface
Modulus, E (mpsi)	62.0	12.3	12.3
Poisson's ratio, ν	0.3	0.32	0.32
Shear Modulus, G (mpsi)	23.8	4.6	4.6
Coeff. of Thermal Expansion α (ppm/ °F)	1.8	4.5	4.5

PROCEDURE :

- For an assumed fracture mode, uniform displacement boundary conditions are applied to the model in a given direction.
- Resulting forces corresponding to those applied displacements are computed by finite element analysis.
- Comparison of resulting forces is made for reduction in global stiffness, and corresponding SERR are computed to identify/quantify the hierarchy of fracture modes.

PROCEDURE (Cont.) :

- In the sample illustration, fracture is initiated at the middle of a fiber and propagated in the fiber-matrix interface

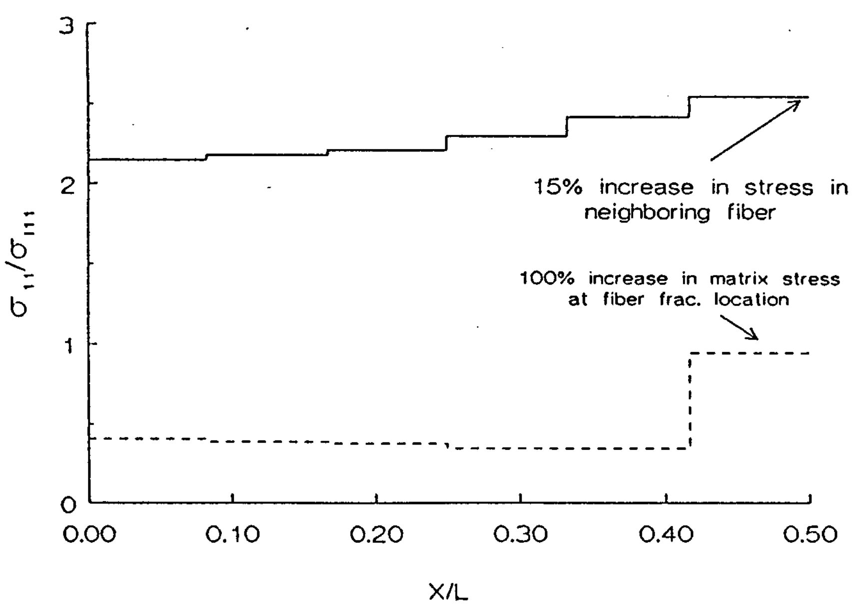


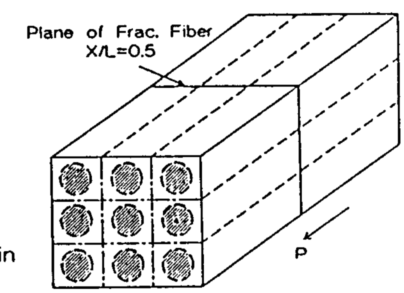
Strain Energy Release Rate (SERR)

- Strain Energy Release Rate (G) gives a measure of the amount of energy required to propagate a defect in the laminate, and to make a direct comparison of damage tolerances between microfracture modes.
  - Crack closure method (local or microfracture approach)
  - Global approach :

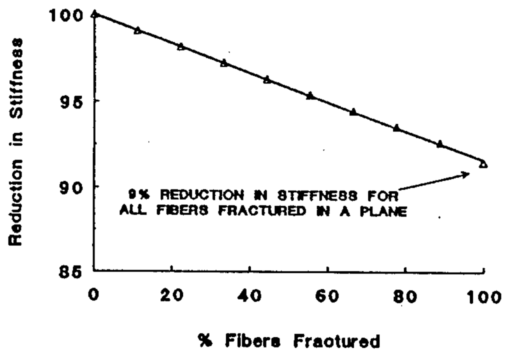
$$G = \frac{dW}{dA} = \frac{1}{2}\frac{F_2 u - F_1 u}{\Delta A}$$

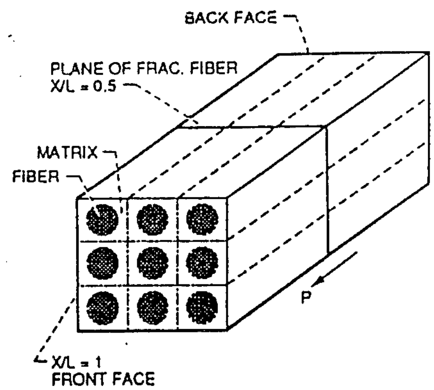
Neighboring Fiber and Matrix Stress Increases Due to Fiber Fracture (Longitudinal Loading); SiC/Ti15, FVR .35

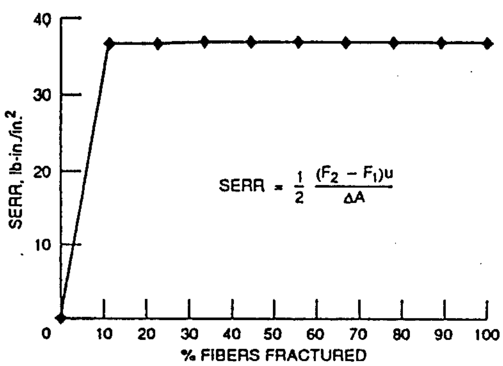




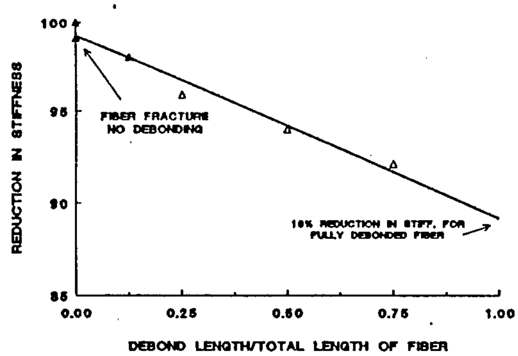
REDUCTION IN LONGITUDINAL STIFFNESS AND CORRESPONDING STRAIN ENERGY RELEASE RATES AS FRACTURE PROPAGATES

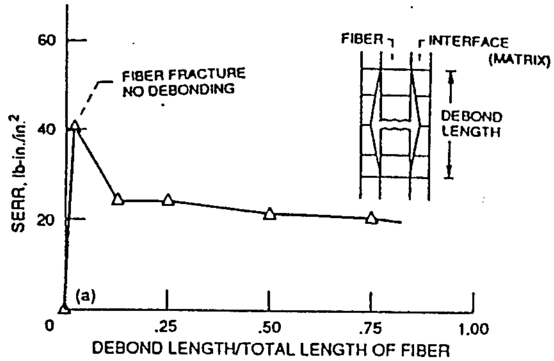


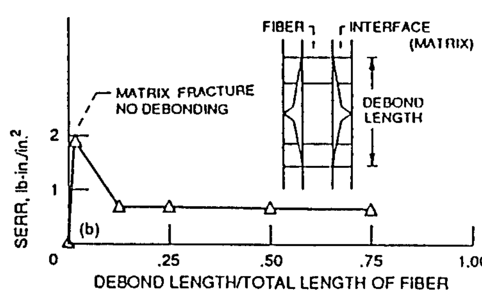




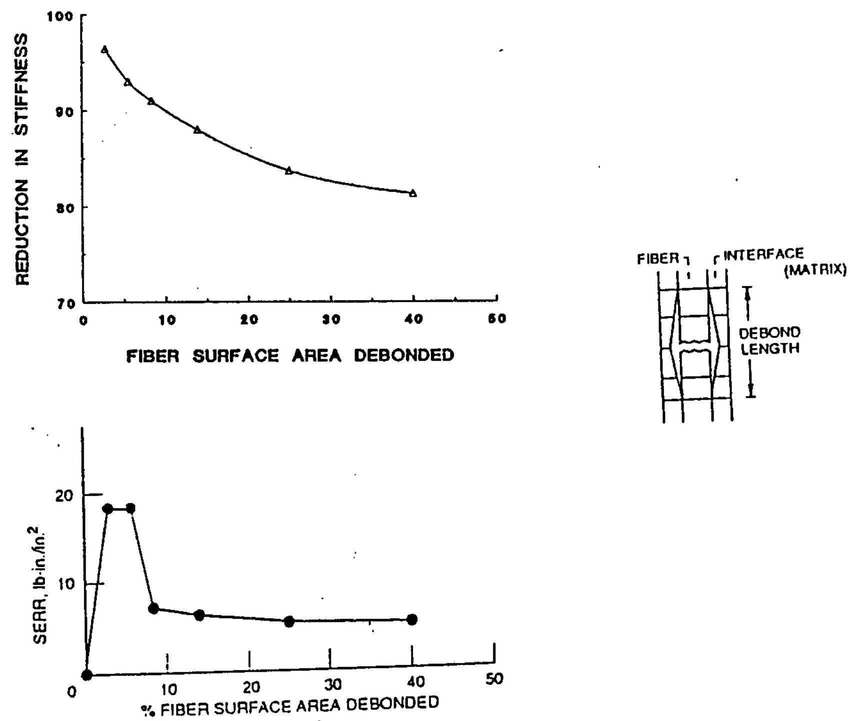
PROCEDURE AIDS IN THE IDENTIFICATION/QUANTIFICATION OF MICROFRACTURE MODE HIERARCHY (LONGITUDINAL LOAD)



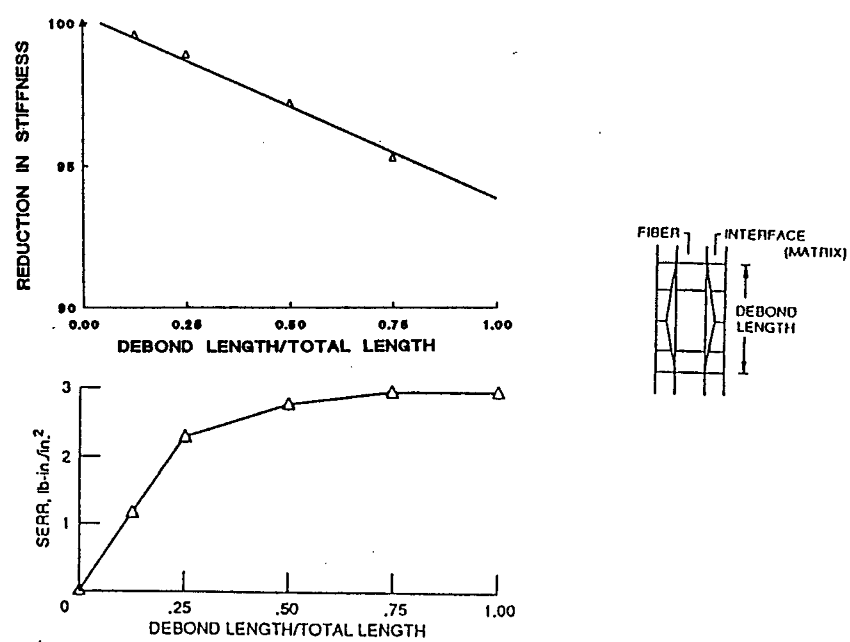




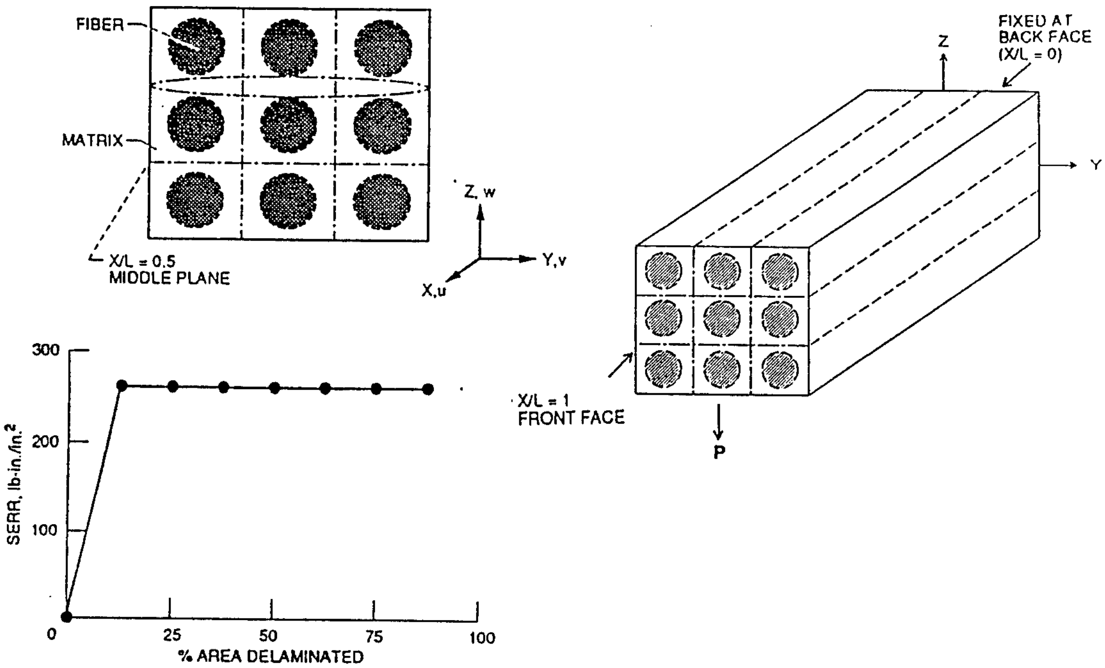
PROCEDURE AIDS IN THE IDENTIFICATION/QUANTIFICATION OF MICROFRACTURE MODE (TRANSVERSE LOAD)



PROCEDURE AIDS IN THE IDENTIFICATION/QUANTIFICATION OF MICROFRACTURE MODE (SHEAR LOAD)



PROCEDURE AIDS IN THE IDENTIFICATION/QUANTIFICATION OF MICROFRACTURE MODE (BENDING LOAD)



Guidelines for Evaluating Composite Microfracture

- Generate a 3-dim. finite element model of the specimen and obtain room temperature properties.
- Perform simulation for ref. (no fracture) case under desired loading condition (longitudinal, transverse etc.)
- Initiate and propagate the fracture under same loading condition as in step 2 and perform various computational simulations for different extent of damage.
- Compare the different fracture configurations for reduction in global stiffness and corresponding SERR.
- Determine the most likely direction for crack propagation based on SERR. It may happen that a certain configuration requiring high fracture energy has to be reached prior to a configuration requiring low fracture energy.

CONCLUSIONS

- Stresses redistribute around a microfracture within a short distance.

- Longitudinal Load :
  - Fracture in one fiber is unlikely to initiate fracture in neighboring fibers.
  - Interface debonding does not initiate by itself, it follows fiber/matrix fracture.
  - Even if substantial percentage of fibers are fractured in a plane, the reduction in global longitudinal stiffness is small and perhaps difficult to detect experimentally.

- Transverse or Shear Load :
  - Debonding along fiber-matrix interface is the only likely mode of fracture propagation under such loading.

- Bending Load :
  - Interply delamination is the only likely mode of fracture propagation.

A COMPUTATIONAL PROCEDURE TO TAILOR INTERPHASE LAYER CHARACTERISTICS AND PROCESSING HISTORY OF MMCs FOR MINIMUM RESIDUAL STRESSES

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Metal-matrix composites (MMCs) are potential candidate materials for applications demanding high operational temperatures (400 C to 1100 C). Additional high specific moduli and strengths, tailorable properties, dimensional stability, and hygral resistance makes these materials especially attractive for use in the aerospace industry. However, a crucial problem limiting the use of many MMCs is the high residual thermal microstresses developed during the fabrication process, as a result of the large temperature differential and the mismatch between the thermal expansion coefficients (CTE) of the fiber and matrix. The presence of high residual microstresses in the matrix degrades the mechanical properties and the thermomechanical fatigue endurance of the composite. Possible ways to reduce the thermal residual microstresses appear to be the improvement of the fabrication process and the use of an interphase layer between the fibers and the matrix with compatible thermal and mechanical properties [1-2]. The purpose of this paper is to present a computational methodology for the concurrent tailoring of the fabrication process and the interphase of MMCs for minimal residual stresses.

The objective is to minimize the residual microstresses at the end of the fabrication process by optimizing the temperature and consolidation pressure profiles, together with the compliant layer properties (modulus, CTE, and strength) and other composite parameters (compliant layer thickness and fiber volume ratio). The in situ integrity of the constituent materials is ensured throughout the process. The thermo-mechanical response of MMCs during their fabrication is simulated based on unified non-linear micromechanics developed in NASA-Lewis Research Center and encoded in METCAN (METal matrix Composite ANalyzer) [3]. The theory incorporates, among other factors, three material phases (fiber, matrix, and interphase), temperature effects, and the nonlinear mechanical behavior of the constituents. The minimization of residual microstresses is formulated as a constrained non-linear mathematical programming (NLP) problem and is numerically solved with the modified feasible directions method.

The proposed methodology is evaluated on the following two MMC systems: (1) ultra-high modulus graphite (P100)/Copper, and (2) silicon carbide (SiC)/titanium 15-3-3-3 (Ti-15). In both cases the fabrication process was optimized first, and then the fabrication process and the interphase were optimized concurrently. The individual fabrication optimization reduced the residual matrix stresses in both composites. The predicted optimum processes follow similar patterns in consolidation temperatures and pressures for both composite systems, that is, the optimal consolidation pressure gradually increases as the consolidation temperature drops, reaching significantly higher values than the pressure of the currently used processes, and then it decreases to zero immediately after the reduction of the temperature to the room value. In this manner, the temperature drop takes place when the pressure is high, and the thermal stresses are forced to develop when the matrix is highly nonlinear and the thermal strains do not induce high stresses. This illustrates the importance of consolidation pressure.

The concurrent fabrication-interphase optimization produced further reductions in the residual stresses of the matrix. For both composites the more crucial interphase property seems to be the coefficient of thermal expansion (CTE). Also, the optimal modulus and the strengths differ slightly than the respective matrix values. The corresponding optimal fabrication process in the case of P100/Cu follows the trend

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[2] Research Engineer, Sverdrup Technology Inc.

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previously described. The optimal fabrication process of SiC/Ti-15 follows a different pattern. In conclusion, all case studies indicated significant margins of reduction in the residual stresses of the matrix, moreover, they demonstrated the versatility and effectiveness of the developed methodology.

REFERENCES

1. D. A. Saravanos, P. L. Murthy, and M. Morel, "Optimum Fabrication Process for Unidirectional Metal-Matrix Composites: Computational Simulation," NASA TM 102559, 1990.

2. L. J. Ghosn and B. A. Lerch, "Optimum Interface Properties for Metal Matrix Composites," NASA TM 102295, 1989.

3. P. L. Murthy, D. A. Hopkins and C. C. Chamis, "Metal Matrix Composite Micromechanics: In-Situ Behavior Influence on Composite Properties," NASA TM 102302, 1989.

A COMPUTATIONAL PROCEDURE TO TAILOR INTERPHASE LAYER CHARACTERISTICS AND PROCESSING HISTORY OF MMCs FOR MINIMUM RESIDUAL MICROSTRESSES

by

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August 1990

OUTLINE

- o BACKGROUND
- o OBJECTIVES
- o THERMOMECHANICAL RESPONSE
- o FABRICATION-INTERPHASE OPTIMIZATION
- o METHOD EVALUATION
- o FUTURE WORK
- o SUMMARY

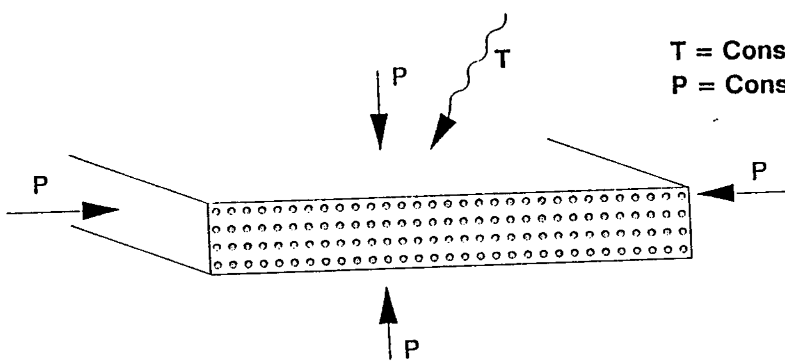
BACKGROUND

- METAL MATRIX COMPOSITES (MMCs) ARE PROMISING HIGH-PERFORMANCE MATERIALS
  - Advanced and tailorable properties
  - High-temperature applications (70-1500 degs F)
  - Dimensional stability
  - Hygral resistance
  - High transverse and shear modulus-strength

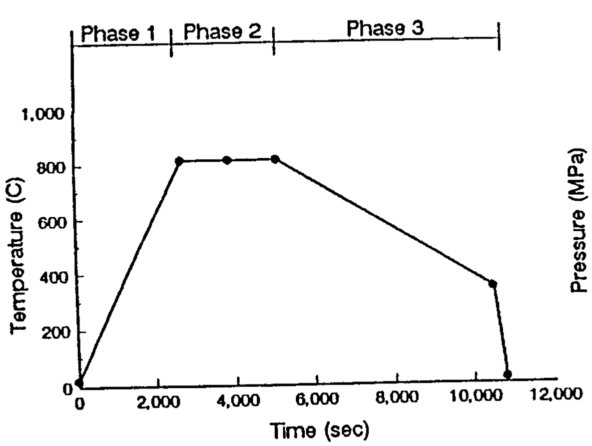
- AN UNRESOLVED PROBLEM REMAINS THE DEVELOPMENT OF HIGH THERMAL RESIDUAL STRESSES DURING FABRICATION
  - Fiber/matrix CTE mismatch
  - High temperature drop

- THE RESIDUAL STRESSES HAVE ADVERSE EFFECTS ON:
  - Mechanical properties of the matrix
  - Thermomechanical fatigue life
  - Material integrity

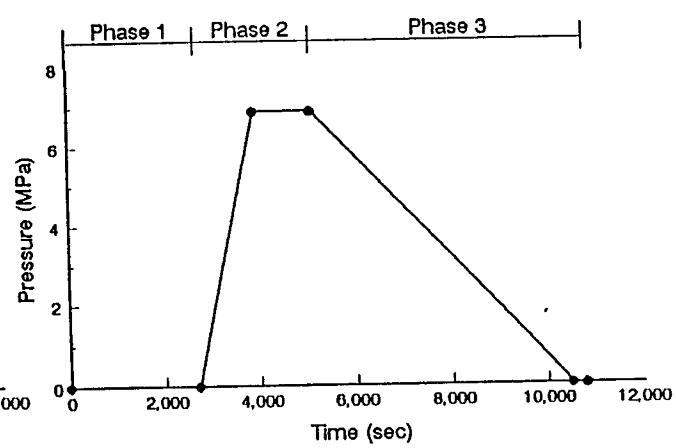
TYPICAL PROCESSING CYCLE FOR GRAPHITE/COPPER COMPOSITE



T = Consolidation Temperature
P = Consolidation Pressure



(a) Temperature



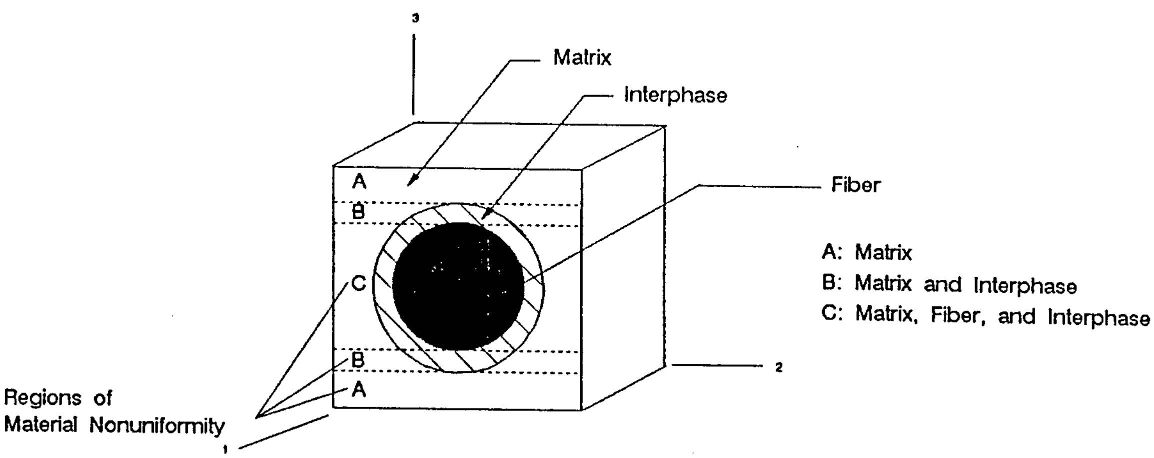
(b) Consolidation Pressure

OBJECTIVES

- o INVESTIGATE THE POSSIBILITY TO REDUCE THE RESIDUAL MICROSTRESSES BY IMPROVING THE FABRICATION PROCESS AND/OR USING A COMPATIBLE INTERPHASE LAYER BETWEEN THE FIBERS AND THE MATRIX.

- o DEVELOPMENT OF A COMPUTATIONAL PROCEDURE TO CONCURRENTLY TAILOR:
  - The fabrication process (consolidation pressure, temperature)
  - The characteristics of the candidate interphase (modulus, CTE, strength, thickness)

- o METHODOLOGY COMBINES FORMAL OPTIMIZATION TECHNIQUES (NON-LINEAR PROGRAMMING) AND METAL-MATRIX COMPOSITE MECHANICS (METCAN)



THERMOMECHANICAL RESPONSE

- o INCREMENTAL NONLINEAR MICROMECHANICS (METCAN) ARE USED TO SIMULATE THE THERMOMECHANICAL RESPONSE OF THE COMPOSITE DURING FABRICATION

- o TEMPERATURE AND STRESS NONLINEAR EFFECTS ON CONSTITUENTS ARE INCLUDED

$$\frac{P_i^t}{P_{oi}} = \left[\frac{T_{Mi} - T^t}{T_{Mi} - T_o}\right]^q \left[\frac{S_i^t - \sigma_i^t}{S_i^t}\right]^p \qquad i = m, d, f$$

- o PROVISIONS FOR THREE MATERIAL PHASES (FIBER, INTERPHASE, MATRIX)

- o CLOSED-FORM EXPRESSIONS FOR COMPOSITE PROPERTIES AND MICROSTRESSES AT THREE MICROREGIONS

FABRICATION-INTERPHASE OPTIMIZATION

OBJECTIVE FUNCTION:

- Minimize the maximum residual microstresses in the matrix

$$\min(\max\{ \sigma_{mA11} + \sigma_{mA22} \})$$

CONSTRAINTS:

- Upper-lower bounds on optimization variables (Fabrication - Interphase parameters)
- Stress failure constraints on matrix, interphase, and fiber stresses

$$S^t_{Cij} < \sigma^t_{ij} < S^t_{Tij}$$

where:
S^t_C = compressive strength at time step t
S^t_T = tensile strength at time step t
σ^t = stress at time step t
i,j = direction 1,2,3

- Admissible interphase thickness

METHOD EVALUATION

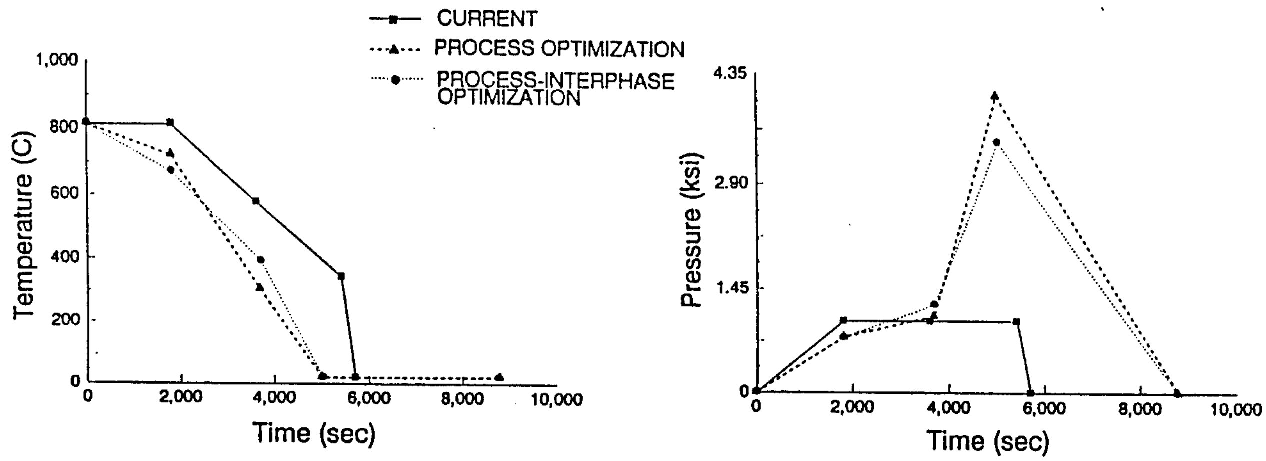
o COMPOSITE SYSTEMS:

(1) Ultra-high modulus Graphite/Copper (P100/Cu)
(2) Silicon-Carbide/Titanium-15 (SiC/Ti-15)

o OPTIMIZATION CASES:

(a) Fabrication process tailoring only (w/o interphase)
(b) Concurrent fabrication-interphase optimization

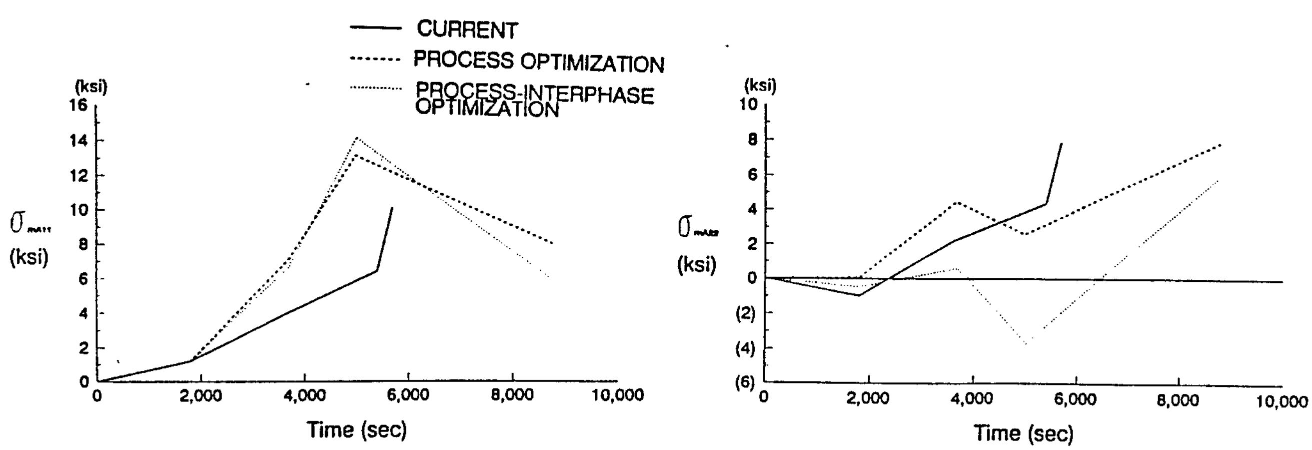
OPTIMUM FABRICATION PROCESSES
P100/COPPER



OPTIMUM INTERPHASE CHARACTERISTICS
P100/COPPER

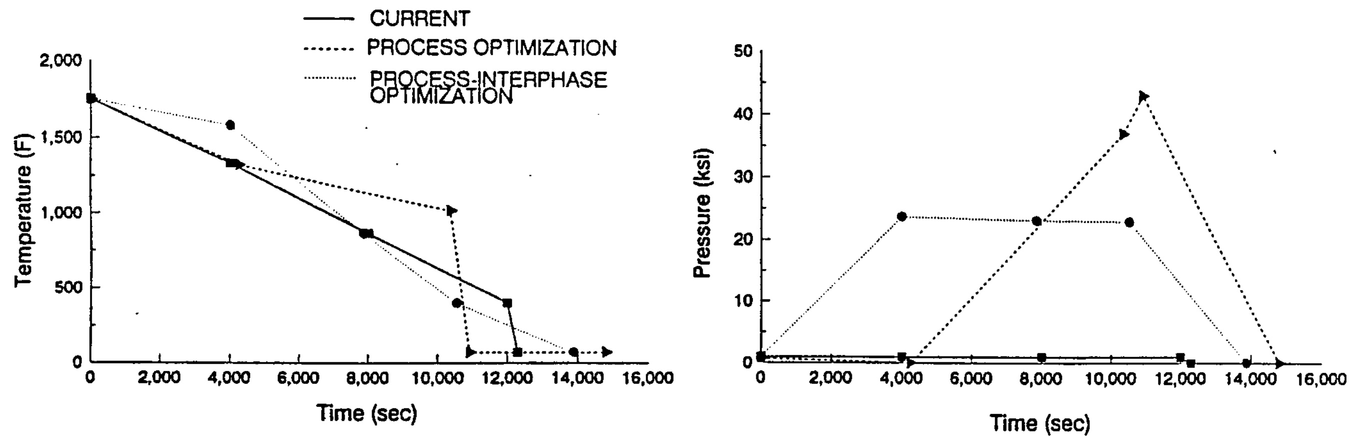
Initial (Matrix)	Optimum
$E_d = 17.7$ Mpsi	$E_d = 18.8$ Mpsi
$\alpha_d = 9.80\ \mu in/in/^\circ F$	$\alpha_d = 15.0\ \mu in/in/^\circ F$
$S_d = 32.0\ ksi$	$S_d = 27.0\ ksi$
$k_d = 12\%$	$k_d = 15\%$
$k_f = 40\%$	$k_f = 47\%$

MATRIX MICROSTRESSES*
P100/COPPER



(*) Stresses at the end of the process are the residual stresses

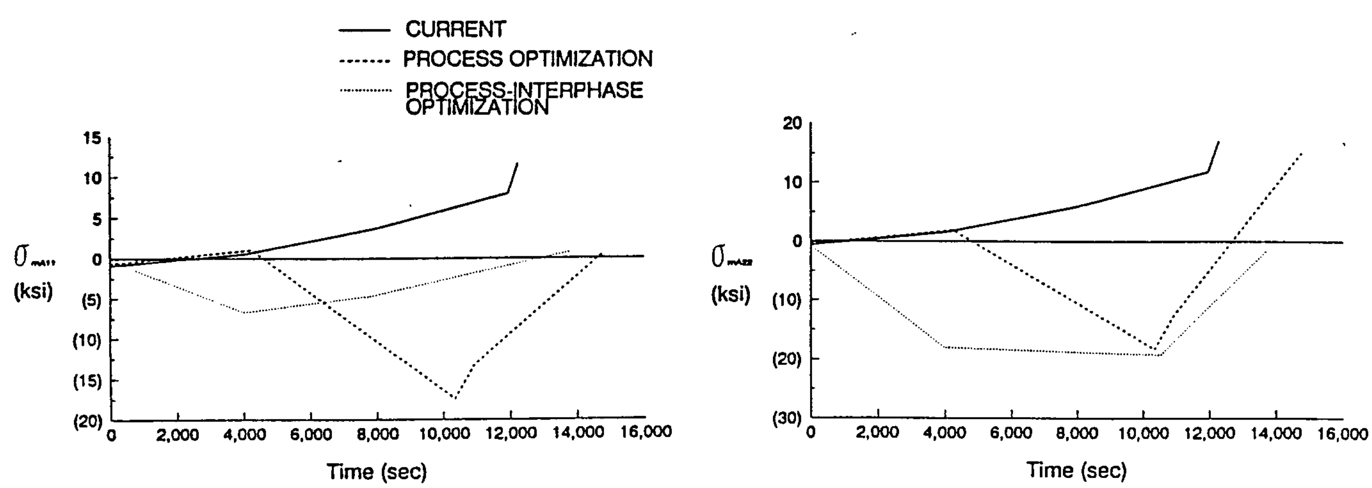
OPTIMUM FABRICATION PROCESSES
SiC/Ti-15-3



OPTIMUM INTERPHASE CHARACTERISTICS
SiC/Ti-15-3

Initial (Matrix)	Optimum
$E_d = 12.3$ Mpsi	$E_d = 12.8$ Mksi
$\alpha_d = 4.50\ \mu in/in/^{\circ}F$	$\alpha_d = 7.80\ \mu in/in/^{\circ}F$
$S_d = 130.0$ Mksi	$S_d = 129.7\ ksi$
$k_d = 12\%$	$k_d = 13\%$
$k_f = 40\%$	$k_f = 23\%$

MATRIX MICROSTRESSES*
SiC/Ti-15-3



(*) Stresses at the end of the process are the residual stresses

FUTURE WORK

- EXTENSION OF THE METHODOLOGY TO COMPOSITE LAMINATES
- ADDRESS THE CONCEPT OF CONCURRENT FABRICATION AND LAMINATE TAILORING FOR MAXIMUM ENDURANCE AT SPECIFIC THERMOMECHANICAL CONDITIONS

SUMMARY

- A COMPUTATIONAL PROCEDURE TO TAILOR THE INTERPHASE AND THE PROCESSING HISTORY OF MMCs WAS DEVELOPED
- NON-LINEAR COMPOSITE MECHANICS WERE USED TO SIMULATE THE THERMOMECHANICAL RESPONSE OF THE COMPOSITE DURING FABRICATION
- APPLICATIONS OF THE METHOD ON GRAPHITE/COPPER AND SIC/TI-15-3 COMPOSITES DEMONSTRATED THE FEASIBILITY OF THE CONCEPT.
- TAILORING OF THE FABRICATION DECREASED THE RESIDUAL STRESSES IN BOTH COMPOSITES
- CONCURRENT OPTIMIZATION OF INTERPHASE-FABRICATION RESULTED IN ADDITIONAL SIGNIFICANT REDUCTIONS OF THE RESIDUAL STRESSES
- CONSOLIDATION PRESSURE AND INTERPHASE CTE WERE THE MORE CRUCIAL PARAMETERS
- OVERALL, THE APPLICATIONS ILLUSTRATED THE VERSATILITY AND EFFECTIVENESS OF THE DEVELOPED METHODOLOGY

ENGINEERED MULTIMATERIALS - THE ROLE OF MECHANICS

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ABSTRACT

The art of designing materials has evolved to the point where we are now routinely combining selected constituents to produce multiphase materials with tailored microstructures for specific applications. This trend, already carrying significant momentum, considers the virtually boundless possibilities of constituent combinations and processing options to produce materials which respond to a set of service-environment requirements: higher temperature, enhanced oxidation resistance, higher strength-to-weight and thrust-to-weight ratios, enhanced toughness, etc. Such tailoring can take place over a wide range of dimensional scales and invokes principles from a number of disciplines: chemistry, materials science, solid state physics, mathematics, and mechanics are principal.

This presentation addresses the potential contribution that the mechanics community can make to this emerging approach of engineering materials to meet specific future operational needs. Future requirements for multimaterials naturally invite collaboration and cross-fertilization of ideas across interdisciplinary boundaries. The mechanics community can and should play a key role in this process. Virtually all classes of materials considered viable candidates for meeting projected user needs are highly anisotropic and heterogeneous. The challenge to the community is multi-faceted. We must invite and foster cooperation between our discipline and our colleagues in materials science, chemistry, etc. We need to invite the applied mathematics community to tackle highly nonlinear systems of governing equations. We need to establish the methodology which will guide the application of mechanics principles to the constituents of multiphase materials, thus placing the microstructure-properties connection on a quantitative basis.

Research in the mechanics of multiphase materials should be directed toward identifying and mathematically modelling the physical events which influence their thermomechanical properties and behavior. Constitutive modelling should connect material properties to the original and evolving microstructural features and their interactions. The ultimate goal of mechanics should be reliable simulation of active physical mechanisms and predictions of expected material behavior which would guide the development of candidate material systems.

Specific research needs addressing issues influencing the behavior of certain classes of materials of interest to the aerospace industry are discussed to underscore the need for an integrated, multidisciplinary approach in developing these materials. Ceramics, ceramic composites, and carbon-carbon composites are addressed.

Air Force Basic Research
Aerospace Sciences

2302/B2 STRUCTURAL DURABILITY

Future *Combat Needs* Define Technology Goals

- Expanded Combat Arena
  - Mach No 4-6$^+$
  - Altitude to 100,000$^+$ ft
  - 65% radius increase
- Sustained Supersonic Performance
  - Double reliability
  - Skin temps to 3,000° F
  - Engine temps to 4,000° F
- Greater Maneuverability
  - Higher load factors
  - Sustained, 9 g's; Instantaneous, 12$^+$ g's
- Stealthy Penetration

2302/B2 STRUCTURAL DURABILITY

Critical *Technology Goals* Motivate Research Thrusts

- Reduce structural weight fraction 2X
  - Increase specific strength 3X
  - Multifunctional materials
- Improve Damage Tolerance and Durability of Emerging Materials 2X
  - Damage should not become critical in two lifetimes
  - Damage economically repairable for two lifetimes
  - Materials to maintain nearly room temperature properties at operating environment

Research Thrusts
STRUCTURES - 2302

Current Emphasis

Mesomechanics (89 Initiative)

Soil Microstructure

Nonlinear Dynamics

Inelastic Behavior of Pavement/Construction Materials

Planned (FY 91)

Wave Propagation (Enhancement)

- Granular/Fractured Materials

Planned (FY 92)

High Temperature Ceramics

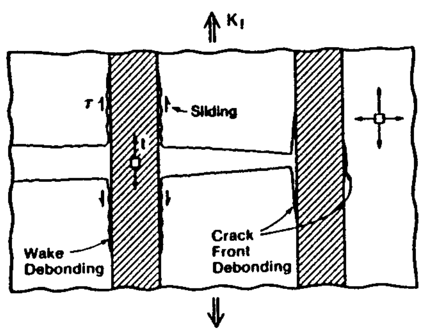
Origin of Imperfections

Carbon-Carbon Composites

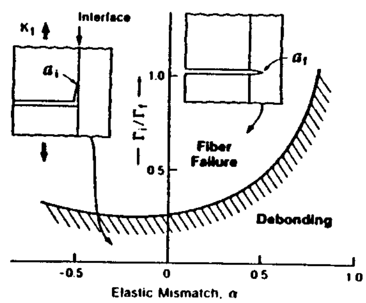
Biomimetics

KNOWLEDGE BASE – TOUGHENING MECHANISMS

FIBER FAILURE SUPPRESSED
AT MATRIX CRACK FRONT



CRITICAL ENERGY RELEASE RATE
REQUIRED FOR CRACK DEBONDING



KNOWLEDGE BASE – INTERFACES IN CMCs

INTERFACE PROPERTIES CONTROL BEHAVIOR OF CMCs

* RATIO OF INTERFACE (DEBOND) TO FIBER TOUGHNESS: $G_{ic}/G_{fc} \leq 1/4$
  - FIBERS DEBOND RATHER THAN BREAK

* RESIDUAL STRAIN MUST BE SMALL ($\Delta\alpha \leq 3 \times 10^{-6}$ /C) AND NEGATIVE
  - PREVENTS THERMAL CRACKING OF FIBERS AND MATRIX
  - INTERFACE IN MODERATE TENSION

* INTERFACIAL COEF OF FRICTION SMALL ($\mu \leq 0.1$)
  - INTERFACE SLIDING RESISTANCE $2 \leq \tau \leq 40$ MPa
  - FIBERS BREAK FAR FROM MATRIX CRACK PLANE (=> LONGER PULLOUT LENGTHS)

AIR FORCE BASIC RESEARCH

FY92 INITIATIVE: HIGH TEMPERATURE BEHAVIOR OF STRUCTURAL CERAMICS

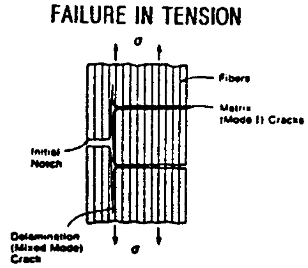
SCIENTIFIC ISSUES AND RESEARCH NEEDS

TOUGHENING OF CMCs

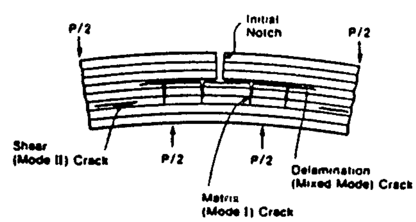
* MECHANICS MODEL WHICH GUIDES THE TAILORING OF THE CONSTITUENTS AND THE INTERFACE FOR SPECIFIC COMPOSITE BEHAVIOR
* IDENTIFY MECHANISMS CONTRIBUTING TO TOUGHENING AND QUANTIFY THE DISSIPATED ENERGY
* ESTABLISH EXPERIMENTALLY AND MODEL ANALYTICALLY THE SEQUENCE, EXTENT AND INTERACTIONS OF THE DAMAGE MECHANISMS
* ANALYTICALLY SIMULATE PROGRESSIVE DAMAGE AND INTERACTIONS, AND PREDICT FAILURE MODES

KNOWLEDGE BASE – FAILURE MODES IN CMCs

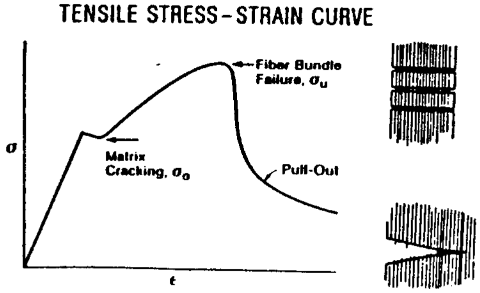
FAILURE IN TENSION



BENDING FAILURE



TENSILE STRESS-STRAIN CURVE



* WEAK INTERFACE
* HIGH-STRENGTH FIBERS
* TENSILE RESIDUAL STRESSES NORMAL TO INTERFACE

AIR FORCE BASIC RESEARCH

FY92 INITIATIVE: HIGH TEMPERATURE BEHAVIOR OF STRUCTURAL CERAMICS

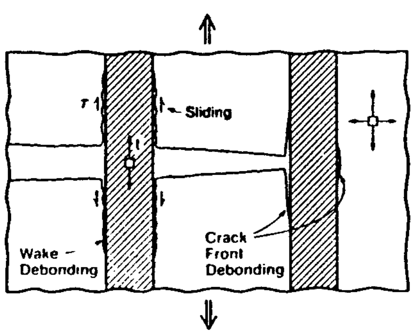
SCIENTIFIC ISSUES AND RESEARCH NEEDS

HIGH TEMPERATURE FATIGUE

* IDENTIFY PHYSICAL MECHANISMS INFLUENCING THE RATE OF GROWTH OF FATIGUE CRACKS
  - USE MICROMECHANICS TO MODEL DEFORMATION, DEGRADATION, AND HYSTERESIS IN PROCESS ZONE NEAR THE CRACK TIP
  - EXPERIMENTALLY VERIFY MODEL PREDICTIONS

* THERMOMECHANICAL FATIGUE IN HOSTILE ENVIRONMENTS
  - MODEL INTERACTION AMONG MECHANICAL CYCLING, THERMAL CYCLING, AND CHEMICAL EVENTS
  - PREDICT SAFE LIFECYCLE ACCOUNTING FOR THE EFFECTS OF OVERLOADS AND UNDERLOADS

BASIC DEBONDING AND SLIDING MECHANISMS



EFFECTS OF TEMPERATURE ON

- DEBONDING
- SLIDING / PULL OUT
- RESIDUAL STRESS

Air Force Basic Research
Aerospace Sciences

FY92 Initiative: High Temperature Behavior of Structural Ceramics*

GOALS

o Identify and quantitatively model damage mechanisms and their interactions

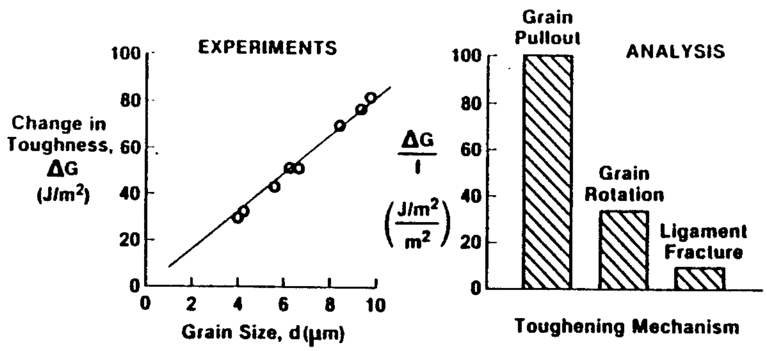
o Guide development of materials tailored for specified performance

o Predict behavior and life times in service conditions

* Pursued jointly with NE and WRDC Materials Laboratory

Air Force Basic Research
Aerospace Sciences

Frictional Grain Pullout found to be the most effective toughening mechanism in monolithic ceramics



Coef. of Friction

- $\Delta G = c\,\mu\,f\,\sigma_{max}\,d$

Fraction of Crack Face Interacting — Grain Size

- $\Delta G = 0.6\,\mu\,f\,\sigma_{max}\,d_{min}$

Transverse Grain Size

- $\Delta G = \frac{1}{18}\,f\,\frac{\sigma_{max}^2}{E}\,d$
- Microcracking Mechanism not present

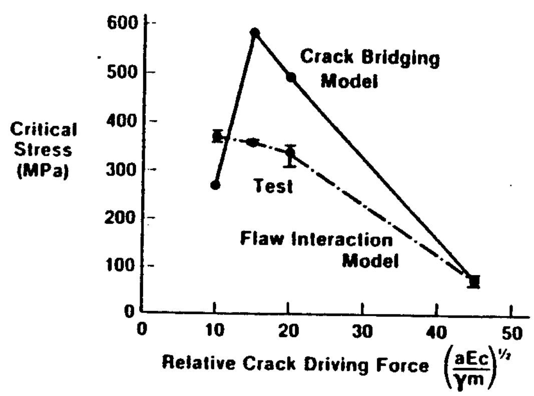
For Tougher Ceramics

- Maximize friction by increasing normal stress between grains
- Include elongated grains with high fracture strength

Ashby, Beaumont/Cambridge

Flaw Interaction Model predicts critical stress for instability in ceramic composites more accurately than Crack Bridging Model

- At Critical Stress matrix cracks suddenly link up with interface flaws



- **Crack Bridging Model**
  - 1-D, Energy Balance
  - Long Crack
  - No Residual Stresses
  - Needs Interface Shear
- **Flaw Interaction Model**
  - 3-D, Griffith Criterion
  - Random Flaws
  - Residual Stresses
  - Interface Shear Not Needed

- Chain of Flaw Interaction: Local Interface Disbonds → Matrix Cracks → Flaw Linking → Critical Damage State

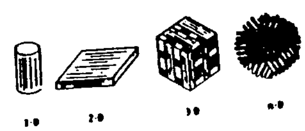
Research

Simulate Flaw Linkage and Define Critical Damage State

Wang/Drexel U.

FUNDAMENTALS OF CARBON/CARBON

BACKGROUND



ULTIMATE TENSILE STRENGTH > 270 MPa (40 KSI)
MODULUS OF ELASTICITY > 69 GPa (10^7 PSI)
MELTING POINT > 4100C
THERMAL CONDUCTIVITY = 11.5 W/mK)
LINEAR THERMAL EXPANSION = 1.1×10^{-6}/C
DENSITY < 2.99 g/cc

C/C LIMITING PROPERTIES ARE THOSE OF GRAPHITE

CRYSTALLOGRAPHIC PROPERTIES CONTROL THE PHYSICAL AND MECHANICAL PROPERTIES

- ANISOTROPY (a vs c)
- DEGREE OF CRYSTALLINITY (AMORPHOUS VS GRAPHITIC)
- PREFEREED ORIENTATION OF THE CRYSTALLITES

FOR EXAMPLE:

ANISOTROPY:

ATOMIC BOND STRENGTH BASAL PLANE = 524 kJ/MOL
c DIRECTION = 7 kJ/MOL

DEGREE OF CRYSTALLINITY:

27 GPa < FIBER MODULUS < 690 GPa

KNOWLEDGE BASE

THE MECHANICS OF C/C COMPOSITES

- THE DESIGN OF SUCCESSFUL OXIDATION PROTECTION SYSTEMS FOR C/C COMPOSITES MUST SIMULTANEOUSLY CONSIDER BOTH:
  - OXYGEN DIFFUSION AND REACTION RATE CHARACTERISTICS
  - EFFECT OF OXIDATION PROTECTION SYSTEMS ON THE THERMO-MECHANICAL PROPERTIES
- THREE TYPES OF OXIDATION PROTECTION SYSTEMS ARE CURRENTLY UNDER DEVELOPMENT:
  - CERAMIC AND GLASS-BASED COATINGS SYSTEMS FOR THE COMPONENT
  - ADJUSTMENTS TO THE CONSTITUENTS
    - COATED FIBERS
    - MATRIX INHIBITORS
  - JOINT USE OF COATINGS WITH INHIBITED MATRICES
- GAINING THE PRESENT LEVEL OF FUNDAMENTAL UNDERSTANDING IN UNPROTECTED C/C COMPOSITES REQUIRED MECHANICS APPLIED AT SEVERAL SCALES:
  - SUB-MICROMECHANICAL
  - MICROMECHANICAL
  - MACROMECHANICAL
- PREDICTING THE EFFECTS OF CONTEMPLATED OXIDATION PROTECTION SYSTEMS ON THE THERMOMECHANICAL BEHAVIOR OF THE RESULTANT COMPOSITE WILL REQUIRE IDENTICAL TREATMENT
  - ADJUSTMENTS TO THE ATOMIC STRUCTURE -> SUB-MICROSCALE
  - EFFECT OF FIBER COATINGS, CRACKS, ETC. -> MICROSCALE
  - PERFORMANCE OF COATED SUBSTRATE -> MACROSCALE

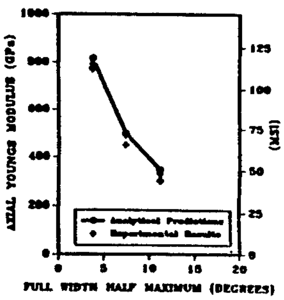
FUNDAMENTALS OF CARBON/CARBON

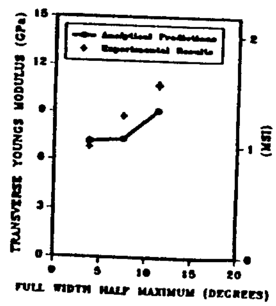
KNOWLEDGE BASE

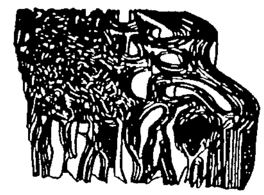
THE MECHANICS OF C/C COMPOSITES
(OR, WHAT IS GEORGE DOING HERE?)

DEMONSTRATED PREDICTIVE CAPABILITY

SUB-MICROMECHANICAL MODEL ELASTIC PROPERTY PREDICTIONS
PITCH FIBERS / UNPROTECTED C/C





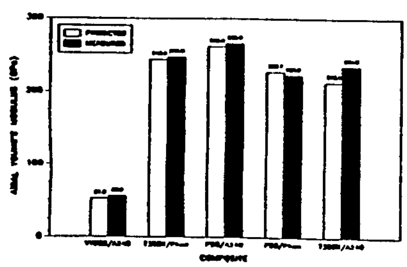


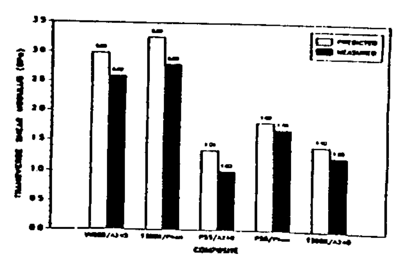
KNOWLEDGE BASE

THE MECHANICS OF C/C COMPOSITES

DEMONSTRATED PREDICTIVE CAPABILITY

MICROMECHANICAL MODEL PREDICTIONS OF UNIDIRECTIONAL
C/C COMPOSITE PROPERTIES





THE MECHANICS OF C/C COMPOSITES

THE ULTIMATE MECHANICS RESEARCH GOAL IS TO DEVELOP THE REQUIRED ANALYTICAL CAPABILITY FOR GUIDING THE DEVELOPMENT OF OXIDATION PROTECTION SYSTEMS FOR STRUCTURALLY USEFUL C/C COMPOSITES

AFOSR FY 92 INITIATIVE - BIOMIMETICS

BACKGROUND & MOTIVATION

- VIRTUALLY ALL STRUCTURAL BIOLOGICAL MATERIALS ARE COMPOSITES
- NATURAL STRUCTURES ARE FUNCTIONAL AND ADAPTIVE
- NATURE BUILDS STRONG, TOUGH STRUCTURES FROM WEAK, BRITTLE INGREDIENTS

• • CALCIUM CARBONATE	STRENGTH:
MONOLITHIC	15-25 MPa
IN NACRE SHELL	185 ± 20 MPa

NATURE'S SECRETS LIE IN COMPOSITE DESIGN AND CRYSTAL GROWTH

A PHYSICAL MODEL OF NACRE

CONSTITUENTS: 95% $CaCO_3$ (CHALK) AND 5% ORGANIC GLUE

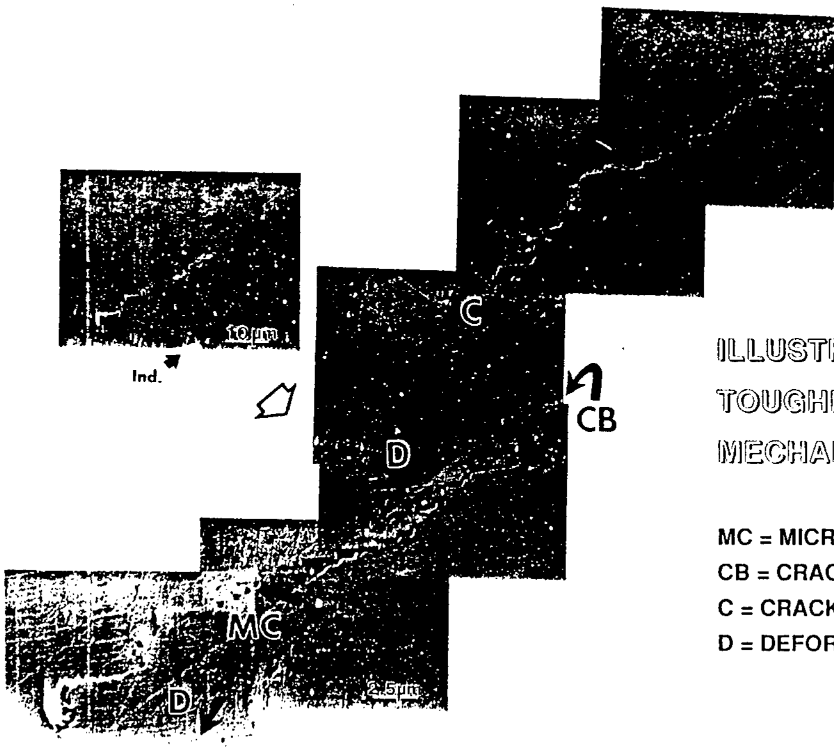
STRUCTURE: "BRICK AND MORTAR"
$CaCO_3$ BRICKS (0.5 μm THICK)
ORGANIC MORTAR (20 - 30 nm)

PROPERTIES: FRACTURE STRENGTH, σ_F = 185 ± 20 MPa
FRACTURE TOUGHNESS, K_{IC} = 8 ± 3 MPa $\sqrt{m}$

COMPARES FAVORABLY WITH MOST "HIGH-TECH" CERAMICS

MANUFACTURERS: SHELLFISH & SNAILS, INC.
(NON-UNION, PATENT <u>NOT</u> APPLIED FOR)

AFOSR FY 92 INITIATIVE - BIOMIMETICS



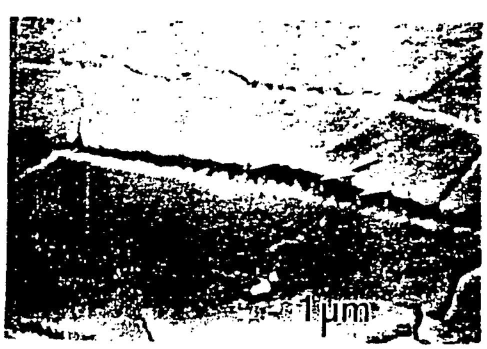
ILLUSTRATED TOUGHENING MECHANISMS

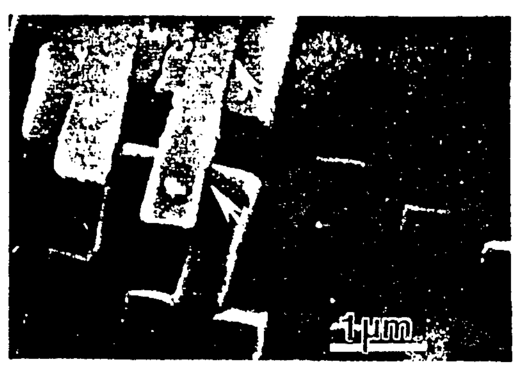
MC = MICROCRACKING
CB = CRACK BLUNTING
C = CRACK BRANCHING
D = DEFORMATION

KEY TOUGHENING MECHANISMS

→ CRACK BRIDGING
- ORGANIC LIGAMENTS STRETCH (UP TO 1000%)
- LOCAL NORMAL STRESS

→ PLATE SLIDING OVER ORGANIC LAYER
- LOCAL SHEAR STRESS

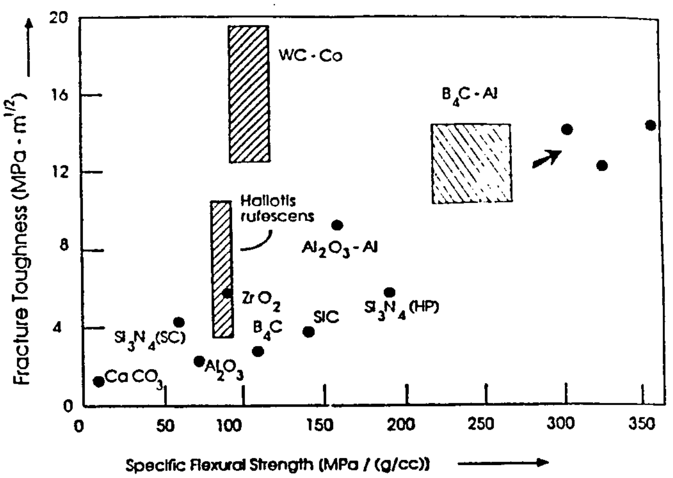




AFOSR FY 92 INITIATIVE - BIOMIMETICS

APPLICATION: BIO - CERMET (B_4C - Al)

- *TOUGHNESS IMPROVED BY 30%*



B_4C - 65% BY VOLUME

Al - 35 % BY VOLUME

LAYER THICKNESS - 10 μm

UNIQUE AIR FORCE BIOMIMETICS GOALS

→ **TECHNOLOGY GOAL:**

PRODUCE AEROSPACE STRUCTURAL MATERIALS WITH SUPERIOR PROPERTIES BY MIMICKING THE PROCESSING AND DESIGN PRINCIPLES MASTERED BY NATURE

→ **BASIC RESEARCH GOAL:**

UNDERSTAND AND DESCRIBE THE STRUCTURE AND FUNCTION OF NATURALLY-EVOLVED MATERIALS

EFFECT OF FIBERS AND TEMPERATURE ON
MATRIX CRACKING IN CERAMIC COMPOSITES

by

F. Delale, Y. L. Xu and B. M. Liaw

The City College of CUNY
Department of Mechanical Engineering
New York, NY 10031

ABSTRACT

The aim of this paper is to present the effect of fiber distribution and temperature on the critical stress intensity factor of a Nicalon/SiC composite. A theoretical model based on finite element computations and as exact analytical solution is presented. It is found that one can accurately predict the experimental results using the model developed in this study.

The material used in the experiments is a Nicalon/SiC ceramic composite. Also monolithic SiC specimens were tested. The composite is a so-called "reversed" composite, meaning that the matrix is stiffer than the fibers.

First the monolithic specimens were tested to ascertain the effect of temperature on the toughness of the SiC matrix. The monolithic specimens consisted of 1/2"x1/2"x1/8" graphite bases coated with a 0.02" SiC layer by the CVD method. The specimens were polished with paste of extra fine alumina or diamond powders (as small as half a micron) and indented with a diamond Vickers indenter at various temperatures from room to 800°C in a series 3320 split-tube laboratory furnace.

For each test the indentation load P was recorded by an ultra precision, 10-lb super-mini load cell, the temperature was read by a type B platinum/platinum- 30% rhodium thermocouple, and the half size of the impression D (as shown in Fig. 3) was measured by a Nikon Um2 universal microscope. From this information one may calculate the critical stress intensity factor of the material using a typical formula as below:

$$K_{Ic} = \frac{1}{\pi^{3/2}\tan\psi}\left(\frac{P}{D^{3/2}}\right), \tag{1}$$

where P is the applied load, D the half crack length and $\psi = 68°$ is the indenter angle. Details of the indentation technique can be found in [1,2].

Next the 1/2"x1/2"x1/8" Nicalon/SiC composite specimens with fibers facing up were also polished as before and in the matrix of the composite, microcracks were generated again using the indentation technique at various temperatures ranging from room to 800°C. To ascertain the effects of fibers, the specimens were indented at locations of varying fiber density and the K_{Ic} was calculated using Eqn. (1), as if the fibers did not exist. Since the K_{Ic} described above contains the effect of fibers, from here on it will be called the "apparent critical stress intensity factor" of the matrix material (SiC). The effect of fibers is introduced through the concept of local volume fraction. The local volume fraction V_f is defined as the ratio of the cross sectional areas of fibers to the total area of the composite cell. It is observed that the "apparent K_{Ic}" decreases with increasing local volume of fibers and also with increasing temperature. This means that it is easier to generate microcracks at a location where the density of fibers is higher and when the temperature is increased.

To explain the observed experimental results, a theoretical model is proposed. It is assumed that the fibers form a ring of composite material with elastic properties calculated based on a volume fraction equivalent to the volume fraction in the cell described above. The theoretical results are obtained through a finite element calculation and a rigorous analytical solution. It is found that the results agree extremely well with the experiments and the prediction from analytical solution is about 2 percent better than the one obtained from the finite element method.

REFERENCES

(1) Evans, A. G., "Fracture Toughness: The Role of Indentation Technique," Fracture Mechanics Applied to Brittle Materials, ASTM STP 678, S. W. Freiman, ed., American Society for Testing and Materials, pp. 112-135., 1979.
(2) Lawn, B. R. and Wilshaw, T. R., "Review - Indentation Fracture: Principle and Applications," Journal of Materials Science, Vol. 10, pp. 1049-1081, 1975.

SPECIMEN OBSERVATION INDICATES MANY FIBER DISTRIBUTION PATTERNS

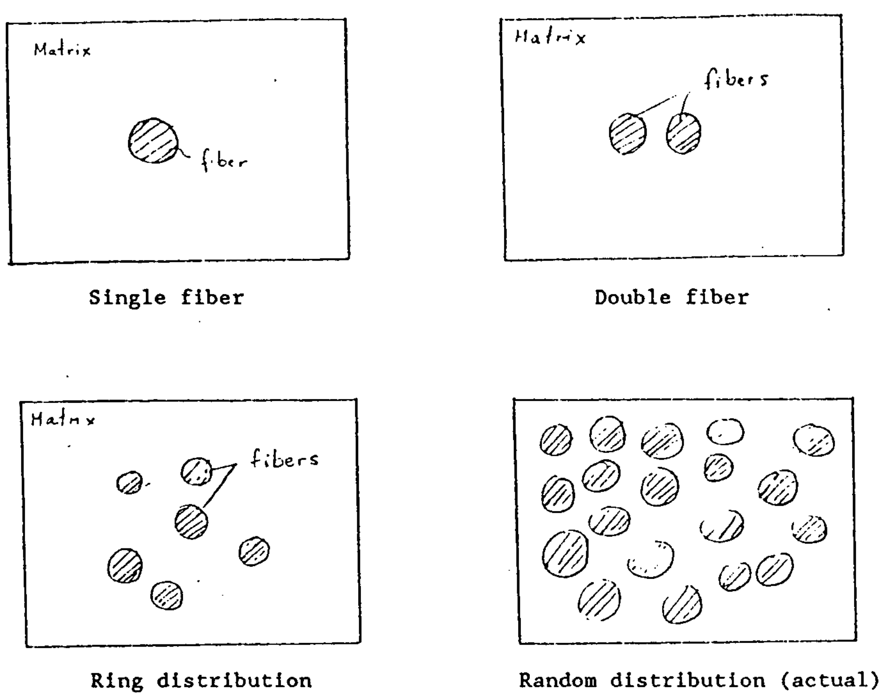


Fig.1 Fiber distribution patterns

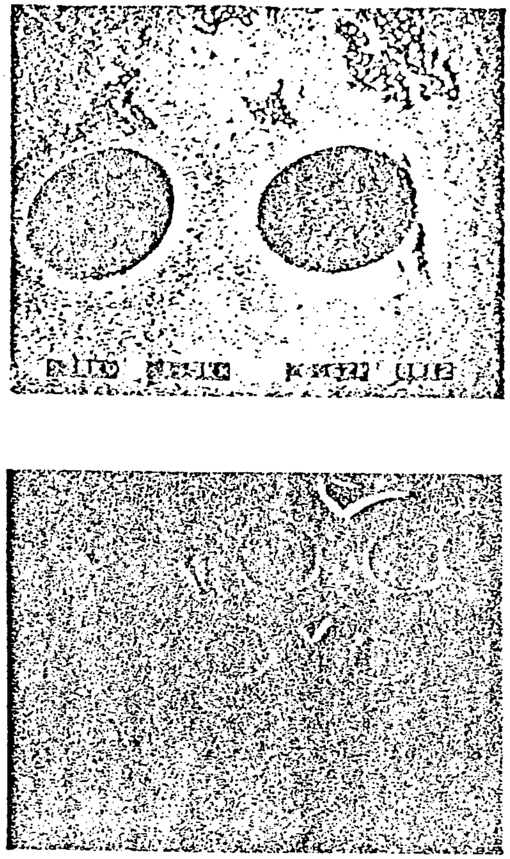
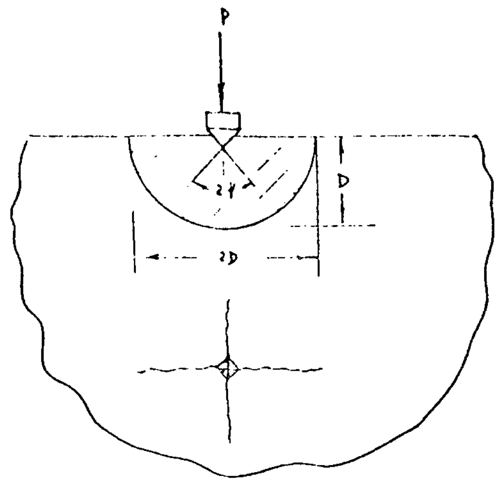


Fig.2 Fibers in ceramic matrix



FORMULA FOR MEASURING K_{IC} FROM

INDENTATION FRACTURE TECHNIQUE

$$K_{IC} = \frac{1}{\pi^{3/2} \tan \psi} \left(\frac{P}{D^{3/2}} \right), \qquad \psi = 68^{\circ}$$

Fig.3 Determination of K_{IC} by micro-indentation technique

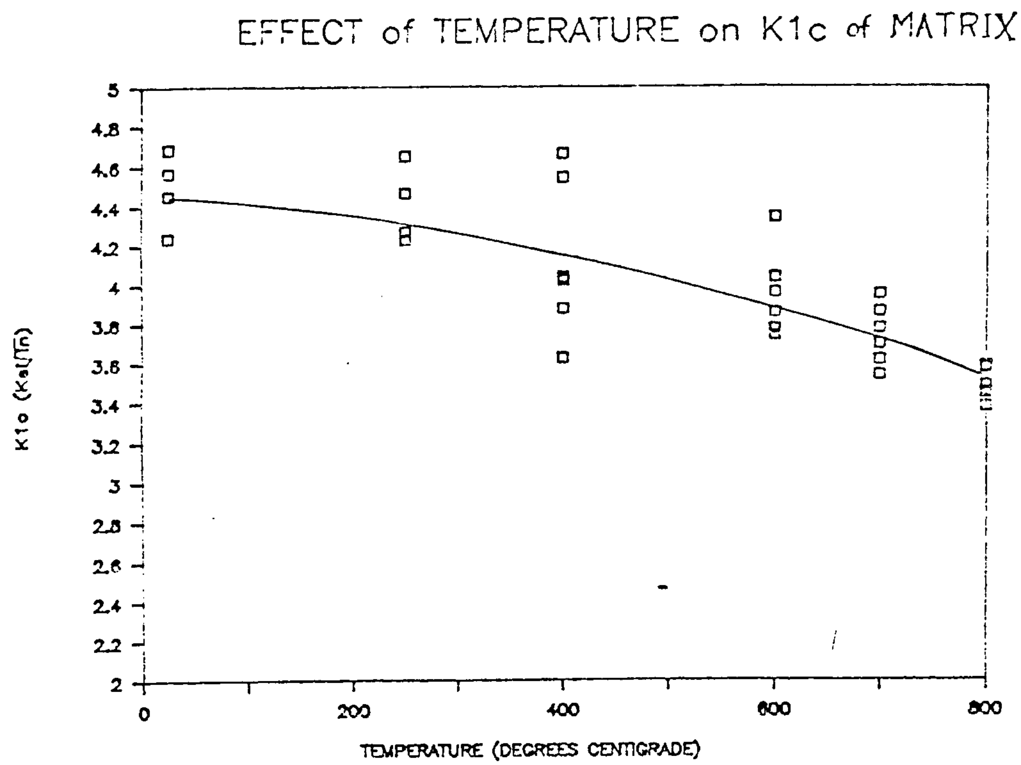

EFFECT of TEMPERATURE on K1c of MATRIX
K1c (Ksi√in)
TEMPERATURE (DEGREES CENTIGRADE)

Fig.4

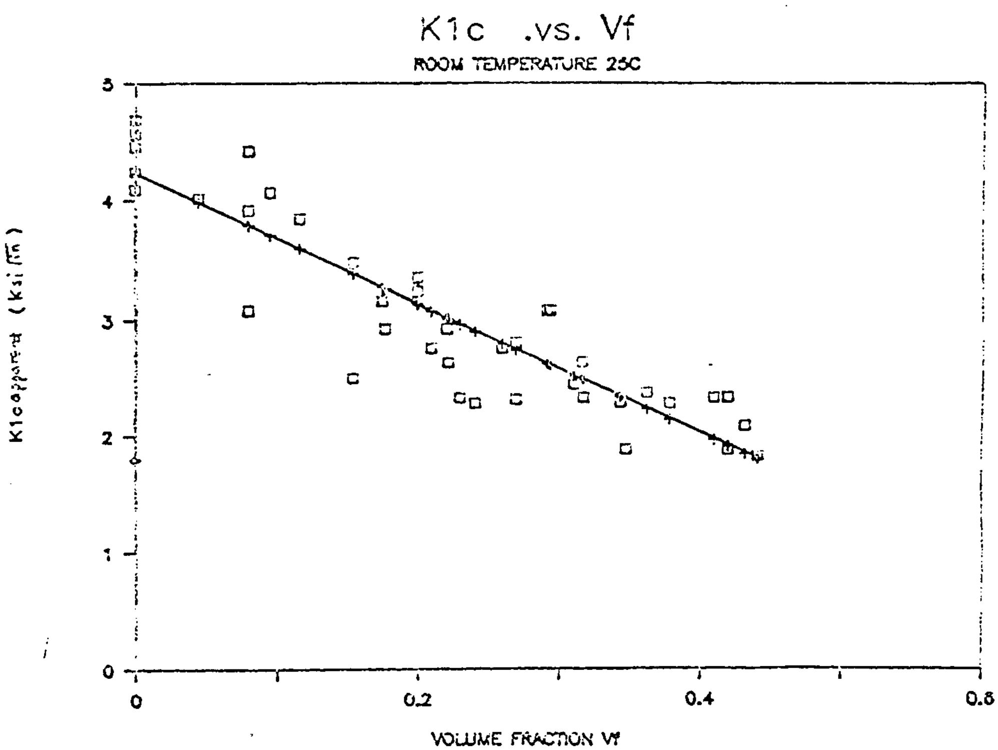

K1c .vs. Vf
ROOM TEMPERATURE 25C
VOLUME FRACTION Vf

Fig.5

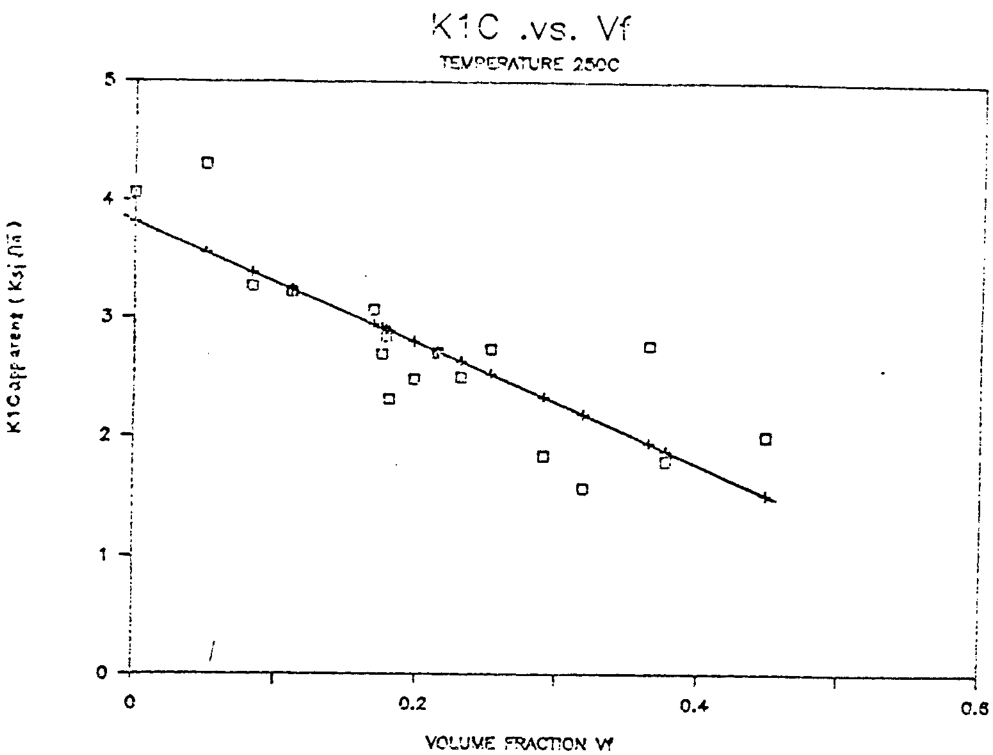

K1C .vs. Vf
TEMPERATURE 250C
K1Capparent (Ksi√in)
VOLUME FRACTION Vf
0
1
2
3
4
5
0.2
0.4
0.6

Fig.6

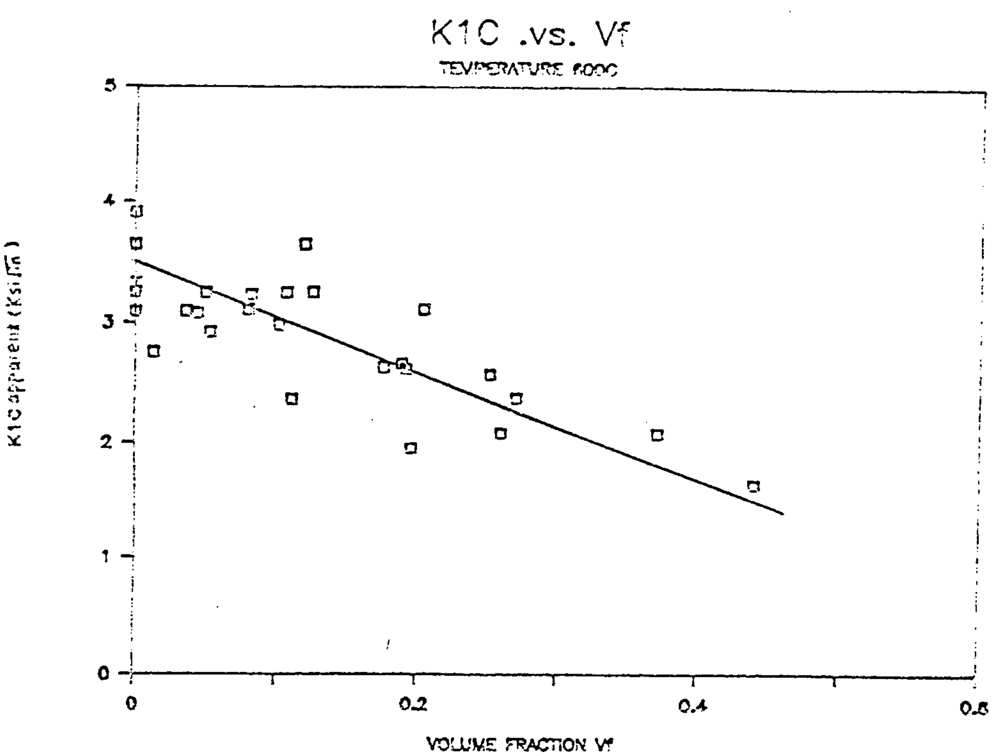

K1C .vs. Vf
TEMPERATURE 600C
K1Capparent (Ksi√in)
VOLUME FRACTION Vf
0
1
2
3
4
5
0.2
0.4
0.6

Fig.7

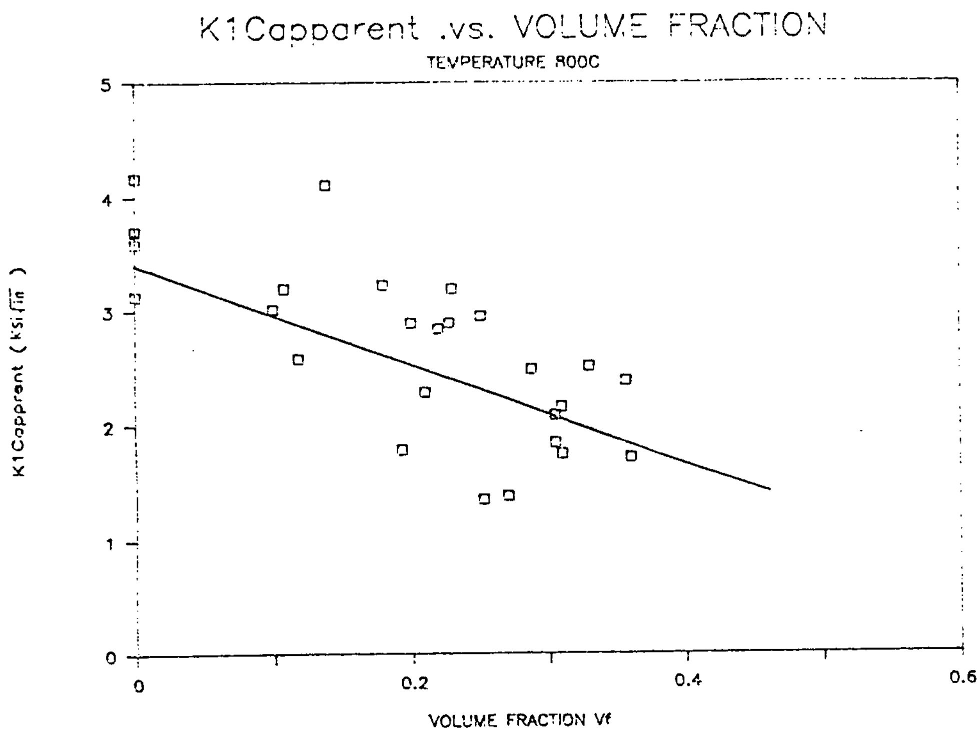

K1Capparent .vs. VOLUME FRACTION
TEMPERATURE 800C
K1Capprent (Ksi√in)
VOLUME FRACTION Vf
0
0.2
0.4
0.6
0
1
2
3
4
5

Fig.8

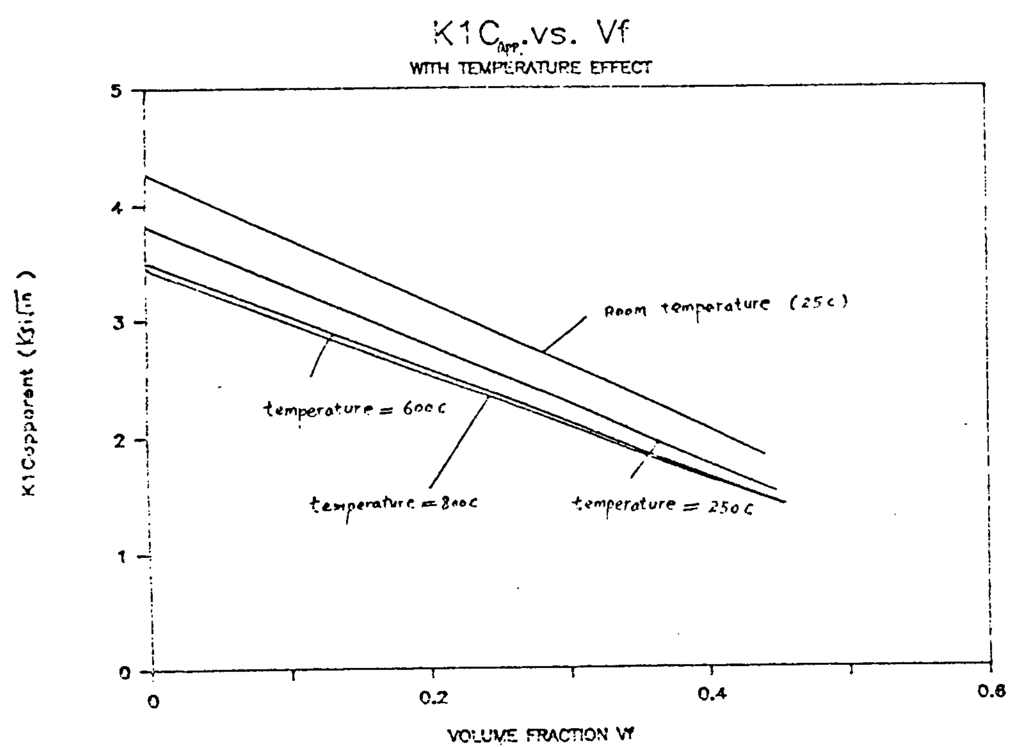
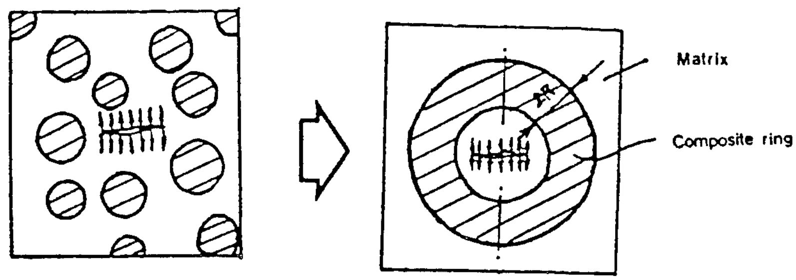

K1C_App. vs. Vf
WITH TEMPERATURE EFFECT
K1Capparent (Ksi√in)
VOLUME FRACTION Vf
Room temperature (25c)
temperature = 600c
temperature = 800c
temperature = 250c
0
0.2
0.4
0.6
0
1
2
3
4
5

Fig.9



(a) actual distribution of fibers

(b) Model

Fig.10

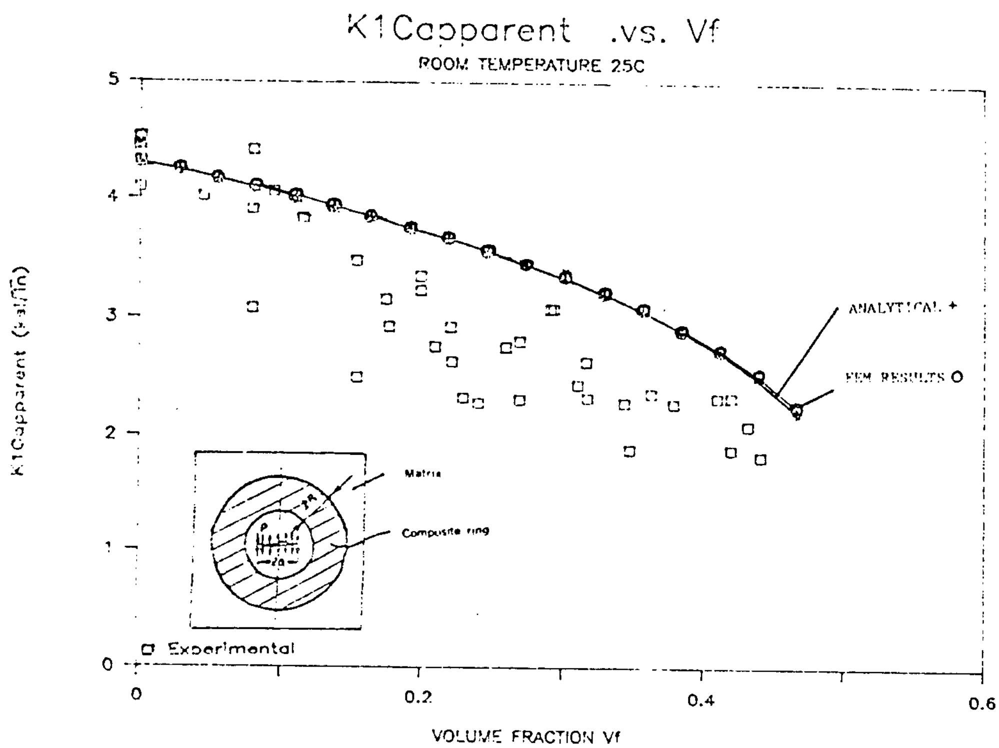


Fig.11

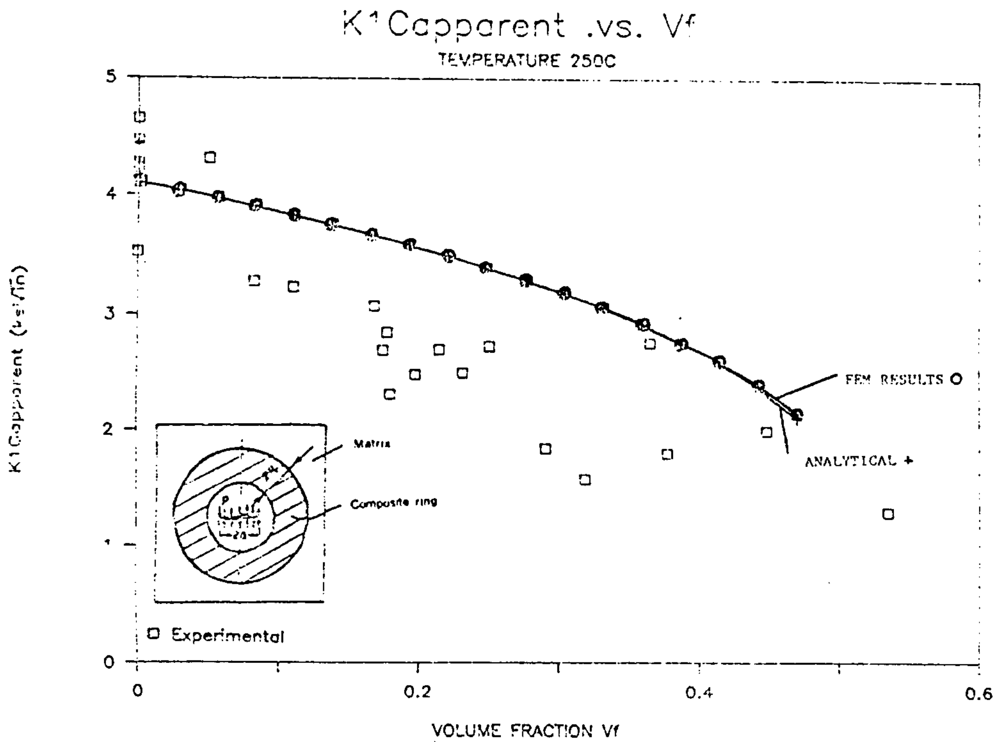

K1Capparent .vs. Vf
TEMPERATURE 250C
K1Capparent (ksi√in)
VOLUME FRACTION Vf
FEM RESULTS O
ANALYTICAL +
Matrix
Composite ring
Experimental

Fig.12

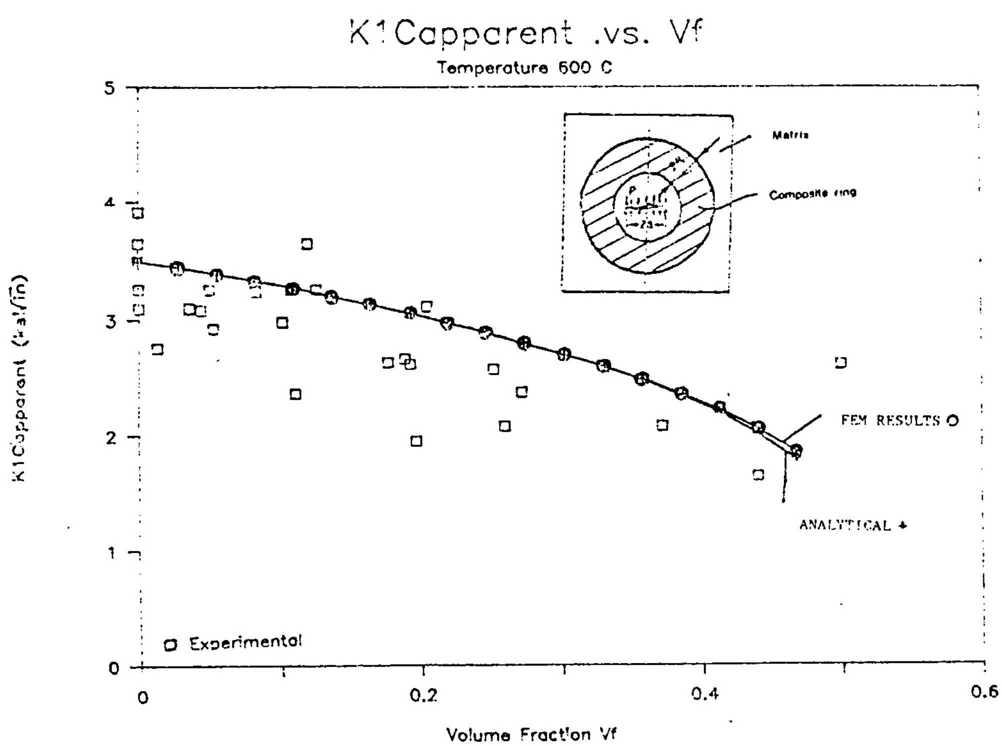

K1Capparent .vs. Vf
Temperature 600 C
K1Capparent (ksi√in)
Volume Fraction Vf
FEM RESULTS O
ANALYTICAL +
Matrix
Composite ring
Experimental

Fig.13

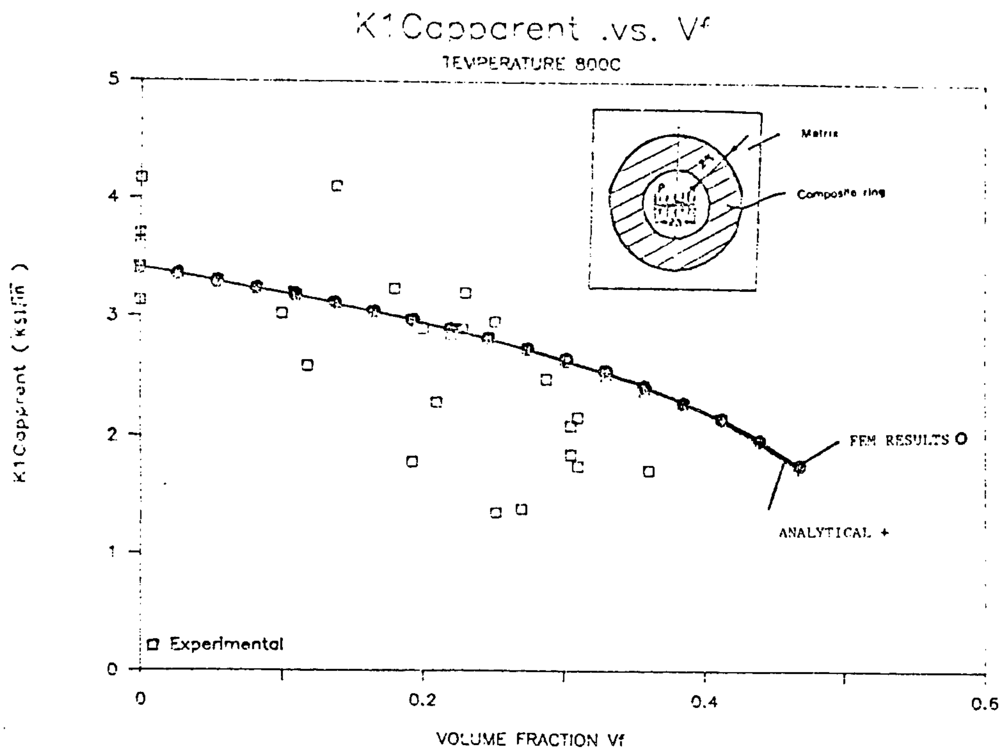
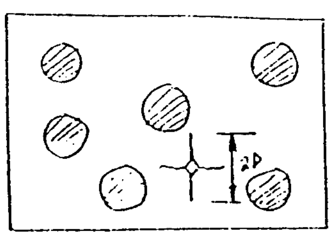


Fig.14

EFFECT OF FIBERS



CONCEPT OF APPARENT FRACTURE TOUGHNESS

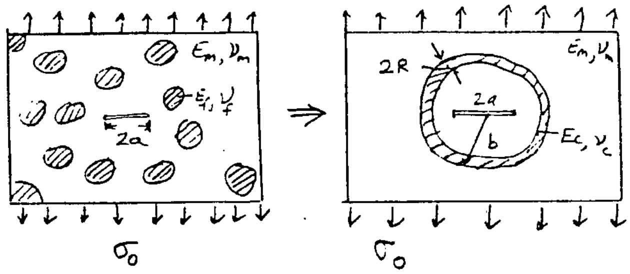
Compute K_{IC} for matrix as if fibers did not exist

$$(K_{IC})_{apparent} = \frac{1}{\pi^{3/2} \tan \psi} \left(\frac{P}{D^{3/2}} \right)$$

$(K_{IC})_{app}$ varies with location (due to effect of neighboring fibers)

We plot $(K_{IC})_{app}$ vs <u>Local volume fraction</u>

THEORETICAL MODEL



$$K = \sigma_0 \sqrt{\pi a}\; f\left(\frac{R}{a}, \frac{b}{a}, n, \frac{E_m}{E_f}, \Delta T\right)$$

or

$$(K_{IC})_m = (\sigma_0)_c \sqrt{\pi a}\; f\left(\frac{R}{a}, \frac{b}{a}, n, \frac{E_m}{E_f}, \Delta T\right)$$

$$(K_{IC})_m = (K_{IC})_{app}\;\; f\left(\frac{R}{a}, \frac{b}{a}, n, \frac{E_m}{E_f}, \Delta T\right)$$

$$= (K_{IC})_{app}\;\; g\left(\frac{R}{a}, V_f, \Delta T\right)$$

or $$(K_{IC})_{app} = \frac{(K_{IC})_m}{g\left(\frac{R}{a}, V_f, \Delta T\right)}$$

g: Normalized stress intensity factor

MICRO-MACRO CONTINUUM MODELING OF COMPOSITE LAMINATES

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ABSTRACT

This paper presents a brief account of some recent developments given in [4] on the continuum modeling of composite laminates. The presentation here focuses on the conceptual aspects of the problem. For detailed derivations of the various expressions the reader is referred to [4].

Advance composite materials for aerospace, structural, power and propulsion application offer significant advantages in terms of efficiency and cost. A widespread and efficient application of composite materials requires detailed and reliable knowledge of their physical properties and, in turn, of their behavior under applied loads. Because of potentially diverse structural and physical variety of reinforced composites, it is neither practical nor economical to rely soley on experimental determination of their properties. Therefore, similar to any other branch of physical sciences, it is desirable to develop a theory (or theories) so that one can analyze, explain, and predict the behavior of composite materials under various in-use loading conditions.

In the last three decades several continuum theories have been proposed as models of elastostatics or elastodynamics of composite materials. In general these theories may be divided into two major categories as follows: a) theories that do not account for the effect of microstructure, and b) theories that consider the behavior of microstructure and try to account for its effect in continuum. It is well known that any continuum theory designed to account for the dynamic response of a composite material must, in some fashion, reflect the effect of microstructure in the composite. In addition to dynamic response of the composite materials, the issue of interlaminar behavior of composite laminates which is directly related to delamination and edge effects in composites is of great importance and it is desirable to somehow incorporate this effect in the formulation of the composite material model(s).

A review of the literature on continuum theories developed for composite laminates reveals that most of the theories that, in some fashion, account for the effect of micro-structure are linear in nature. Moreover, all continuum theories, with the exception of one, are proposed for composite laminates with initially flat configurations. Hence, these theories are not appropriate for curved geometries. In addition the available continuum theories are mainly developed to predict only the dynamic response of the composite laminates. Considering the foregoing, it is quite desirable to develop a continuum theory for laminated composite materials that, if possible, does not suffer from the deficiencies mentioned above. With this in mind a nonlinear continuum theory for laminated composite materials is developed. This theory takes advantage of the theory of Cosserat (directed) surfaces and hence may appropriately be called "*Cosserat composite theory.*"

The composite laminate is modeled as a series of Cosserat surfaces representing the effect of micro-structure. Various field quantities associated with micro- and macro-structures are defined/derived. The theory is represented by a set of well-defined nonlinear and form invariant conservation laws that within the context of purely mechanical theory exhibits the following features: i) it accounts for the effects of microstructures, ii) it accounts for the effects of geometric nonlinearity, iii) it accounts for the interlaminar stresses and therefore delamination can be considered, iv) it is capable of incorporating the effects of material nonlinearity, v) it accounts for the effects of curvature, vi) it possesses a continuum character in the sense of classical continuum mechanics, and finally vii) it is applicable to both static and dynamic problems. The discussion here, due to time and space limitation, is necessarily brief and is confined to conservation laws and the various field quantities appearing in them. Other aspects of the theory such as derivation of constitutive equations, linearized theory, etc., are presented elsewhere [4].

1. Overview

1.a Objective:

To develop a continuum theory for laminated composite materials such that

i) It accounts for the effects of micro-structure
ii) It accounts for the effects of geometric nonlinearity
iii) It accounts for the effects of material nonlinearity
iv) It accounts for the effects of curvature
v) It accounts for the effects of interlaminar stresses
vi) It has a continuum character
vii) It is applicable to both static and dynamic analysis.

1.b Approach:

The above goals will be accomplished by utilizing the following:

i) Convected curvilinear coordinates.
ii) General tensor analysis
iii) Classical three-dimensional continuum mechanics
iv) Theory of Cosserat (directed) surfaces

2. Coordinate Systems

Let the points of a region $\mathcal{R}$ in a three dimensional Eulidean space be referred to a fixed right-handed rectangular Cartesian coordinate system x^i $(i = 1,2,3)$ and let η^i $(i = 1,2,3)$ be a general *convected* curvilinear system defined by the transformation

$$x^i = x^i(\eta^1,\eta^2,\eta^3) \tag{2.1}$$

We assume the above transformation is nonsingular in $\mathcal{R}$ and has a unique inverse

$$\eta^i = \eta^i(x^1,x^2,x^3) \tag{2.2}$$

The existence of the unique inverse implies

$$\det\left(\frac{\partial x^i}{\partial \eta^j}\right) \neq 0 \tag{2.3}$$

For convenience, often we set $\eta^3 = \xi$ and adopt the notation

$$\eta^i = (\eta^\alpha,\xi) \quad , \quad \alpha = 1,2 \tag{2.4}$$

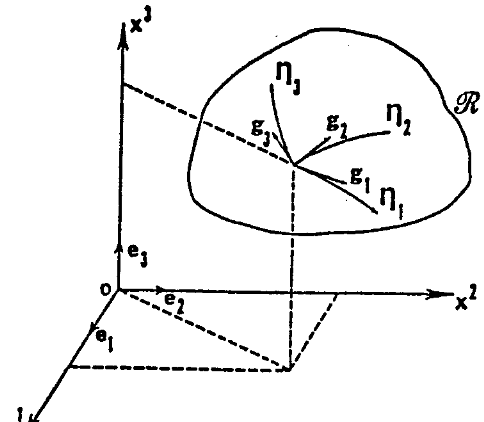

Figure 2-1
Convected curvilinear coordinates

3. Basic field equations of classical continuum mechanics in general curvilinear coordinates

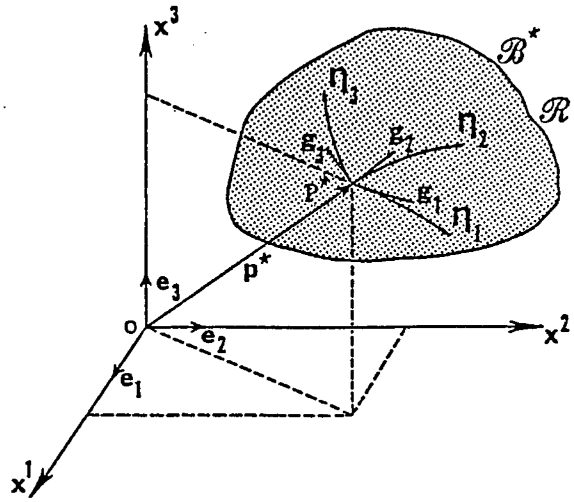

Figure 3-1
A continuum body in the Euclidean 3-space

Let η^i $(i = 1,2,3)$ be a set of general convected curvilinear coordinates. Consider a body $\mathcal{B}^*$ which occupies a region $\mathcal{R}$ in the three dimensional Euclidean space and let its boundary be a closed surface and be denoted by $\partial\mathcal{B}^*$. Let

$$p^* = p^*(\eta^i,t) \tag{3.1}$$

denote the position vector of a material point P^* in the present configuration of the body $\mathcal{B}^*$ at time t. Then we may write

$$g_i^* = \frac{\partial p^*}{\partial \eta^i} \quad , \quad g_{ij}^* = g_i^* \cdot g_j^* \tag{3.2}$$

and

$$ds^2 = dp^* \cdot dp^* = g_{ij}^* d\eta^i d\eta^j \tag{3.3}$$

where (3.2) and (3.3) are the covariant base vectors, the metric tensor, and the square of a line element in the present configuration at time t, respectively. Similarly in the reference configuration we have

$$P^* = P^*(\eta^i) \tag{3.4}$$

$$G_i^* = \frac{\partial P^*}{\partial \eta^i} \quad , \quad G_{ij}^* = G_i^* \cdot G_j^* \tag{3.5}$$

$$dS^2 = dP^* \cdot dP^* = G_{ij}^* d\eta^i d\eta^j \tag{3.6}$$

In addition, we define a strain measure through

$$ds^2 - dS^2 = 2\gamma_{ij}^* d\eta^i d\eta^j \tag{3.7}$$

$$\gamma_{ij}^* = 1/2(g_{ij}^* - G_{ij}^*) \tag{3.8}$$

where γ_{ij}^* are the covariant components of the symmetric strain tensor. Moreover, the velocity is given by

$$v^* = \dot{p}^* = \frac{\partial p^*}{\partial t} \tag{3.9}$$

With reference to the present configuration and within the context of the classical (nonpolar) continuum mechanics, the basic field equations in purely mechanical theory are given by

$$\text{a:}\quad \dot{\rho}^* + \frac{\dot{g}^*}{2g^*}\rho^* = 0$$

$$\text{b:}\quad T^i_{,i} + \rho^* b^* g^{*1/2} = \rho^* \dot{v}^* g^{*1/2}$$

$$\text{c:}\quad g_i \times T^i = 0 \qquad (3.10)$$

$$\text{d:}\quad \rho^* g^{*1/2} \dot{\varepsilon}^* = T^{*i} \cdot v^*_{,i}$$

where we have

$$t^* = \frac{T^{*i} n_i^*}{g^{*1/2}} = \tau^{*ij} n_i^* g_j^* \ ,$$

$$T^{*i} = g^{*1/2} \tau^{*ij} g_j^* = g^{*1/2} \tau_j^{*i} g^{*j} \qquad (3.11)$$

4. Definition of a shell-like body

Consider a body $\mathcal{B}^*$ in the present configuration and let its boundary be a closed surface, denoted by $\partial\mathcal{B}^*$, and composed of three material surfaces as follows:

a) The material surfaces

$$s_1 : \xi = \xi_1(\eta^\alpha)$$

$$\xi_1 < 0 < \xi_2 \qquad (4.1)$$

$$s_2 : \xi = \xi_2(\eta^\alpha)$$

with the material surface

$$s_0 : \xi = 0 \qquad (4.2)$$

lying entirely between them.

b) The material surface

$$s_\ell : f(\eta^\alpha) = 0 \qquad (4.3)$$

such that ξ = const. are closed smooth curves on the surface (4.3), and

c) where the thickness of the shell-like body is small relative to some characteristic length in the body.

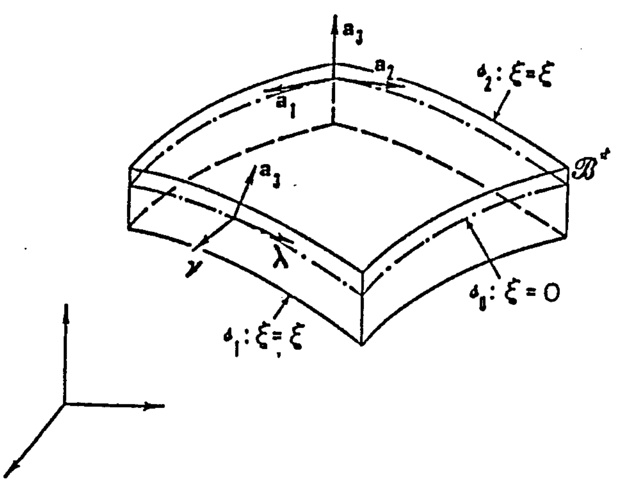

Figure 4-1
A typical shell-like body

5. The basic theory of Cosserat (directed) surfaces

Having introduced the notion of a (three-dimensional) shell-like body, we model such a body by a Cosserat (directed) surface, i.e., a material surface embedded in a Euclidean 3-space, together with a deformable vector field, called *director*, attached to every point of the material surface. The director is not necessarily along the unit normal to the surface and remains unaltered in length under superposed rigid body motions. Let the particles of the material surface, say s, be identified by means of a system of convected coordinates η^α (α = 1,2). Let **r** and **d** denote the position vector of a typical point $\hat{P}$ of s and the director at the same point, respectively.

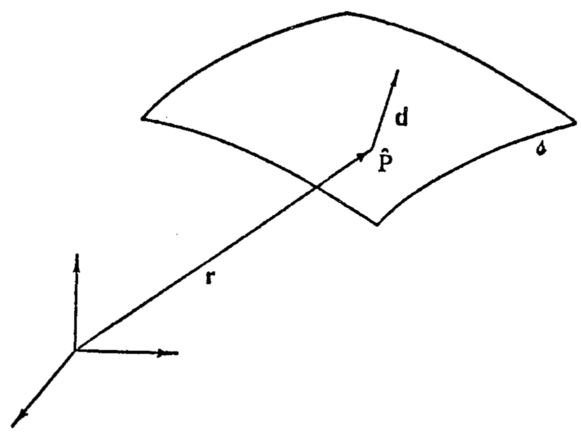

Figure 5-1
A typical Cosserat surface

Then the motion of the Cosserat surface is defined by vector-valued functions which assign position **r** and director **d** to each particle $\hat{P}$ of s at each instant of time through

$$\mathbf{r} = \mathbf{r}(\eta^\alpha, t) \ , \ \mathbf{d} = \mathbf{d}(\eta^\alpha, t) \ , \ [\mathbf{a}_1 \mathbf{a}_2 \mathbf{d}] > 0 \qquad (5.1)$$

where

$$\mathbf{a}_\alpha = \mathbf{a}_\alpha(\eta^\alpha, t) = \frac{\partial \mathbf{r}}{\partial \eta^\alpha} \qquad (5.2)$$

are the base vectors along the η^α-curves on s. The velocity and the director velocity vectors are defined by

$$\mathbf{v} = \dot{\mathbf{r}} \ , \ \mathbf{w} = \dot{\mathbf{d}} \qquad (5.3)$$

With reference to the present configuration, the field equations of a Cosserat surface in the context of purely mechanical theory are given by

$$\text{a} : \ \overline{(\hat{\rho} a^{1/2})}^{\,\cdot} = 0$$

$$\text{b} : \ \hat{\rho} a^{1/2}(\dot{\mathbf{v}} + y^1 \dot{\mathbf{w}}) = (\mathbf{N}^\alpha a^{1/2})_{,\alpha} + \hat{\rho} \hat{\mathbf{f}} a^{1/2}$$

$$\text{c} : \ \hat{\rho} a^{1/2}(y^1 \dot{\mathbf{v}} + y^2 \dot{\mathbf{w}}) = (\mathbf{M}^\alpha a^{1/2})_{,\alpha} - \mathbf{m} a^{1/2} + \hat{\rho} \hat{\mathbf{l}} a^{1/2} \qquad (5.4)$$

$$\text{d} : \ \mathbf{a}_\alpha \times \mathbf{N}^\alpha + \mathbf{d} \times \mathbf{m} + \mathbf{d}_{,\alpha} \times \mathbf{M}^\alpha = 0$$

$$\text{e} : \ \hat{\rho}(\dot{\hat{\varepsilon}}) = \mathbf{N}^\alpha \cdot \mathbf{v}_{,\alpha} + \mathbf{M}^\alpha \cdot \mathbf{w}_{,\alpha} + \mathbf{m} \cdot \mathbf{w}$$

The first of (5.4) is a mathematical statement of the conservation of mass, the second that of the linear momentum, the third is the conservation of the director momentum, the fourth that of the moment of momentum, and the fifth is the conservation of energy. The various quantities appearing in the above conservation laws are defined below:

$\hat{\rho}$: mass density of the surface s

$N = N(\eta^\alpha,t) = N^\alpha(\eta^\alpha,t)\nu_\alpha$: the contact force

$M = M(\eta^\alpha,t) = M^\alpha(\eta^\alpha,t)\nu_\alpha$: the contact director force

where $\nu_\alpha = \nu_\alpha(\eta^\alpha,t)$ are the components of the outward unit normal to the boundary of the shell-like body

$\hat{f} = \hat{f}(\eta^\alpha,t)$: the assigned force

$\hat{l} = \hat{l}(\eta^\alpha,t)$: the assigned director force

$m = m(\eta^\alpha,t)$: the intrinsic director force

$a = a(\eta^\alpha,t)$: determinant of the first fundamental form of the surface.

$y^\alpha = y^\alpha(\eta^\alpha)$: the inertia coefficients

$\hat{\varepsilon} = \hat{\varepsilon}(\eta^\alpha,t)$: the specific internal energy

$\hat{\mathcal{K}} = \hat{\mathcal{K}}(\eta^\alpha,t)$: the kinetic energy of the surface s.

The above quantities are related to the corresponding three-dimensional quantities of the shell-like body and are obtained through suitable integration procedures (for details see [10]).

6. Modeling of a composite laminate as a series of Cosserat (directed) surfaces

We define a composite laminate as a three-dimensional continuum consisting of multiple layers (two or more) of materials which act together as a single (integral) physical entity. Here we confine our attention to laminated composites composed of multiple layers of only two materials, each of which are considered to be homogeneous and isotropic. The layers are not considered to be necessarily flat and could have any type of curvature (see figure 6-1).

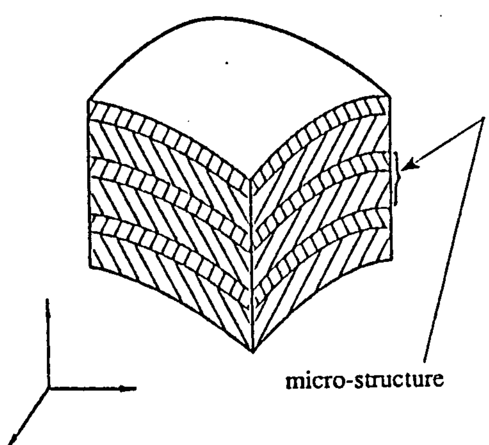

Figure 6-1

A composite laminate consisting of alternating layers of two materials

In order to construct a continuum theory, we should consider a (some) representative feature(s) with repetitive character(s) within the body. For the laminated medium under consideration the most distinct representative feature is the alternating feature of the layers. Hence, we choose the combination of one layer of reinforcement and one layer of matrix as a representative element for the laminated composite. We then model this representative element as a Cosserat (directed) surface using the theory described in the previous section. Next we assume the composite laminate is composed of infinitely many such Cosserat surfaces adjacent to each other. This approach is schematically illustrated below.

Step 1. Selection of a micro-structure

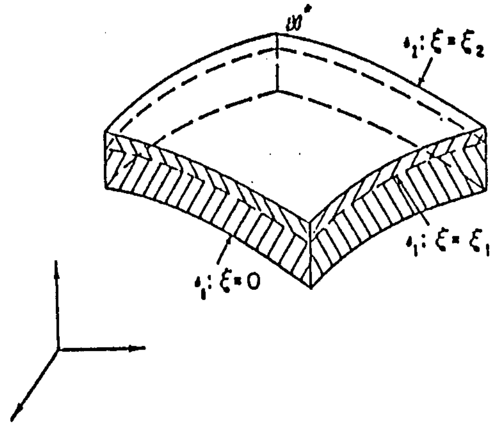

Figure 6-2
Micro-structure

Step 2. Continuum modelling of the micros

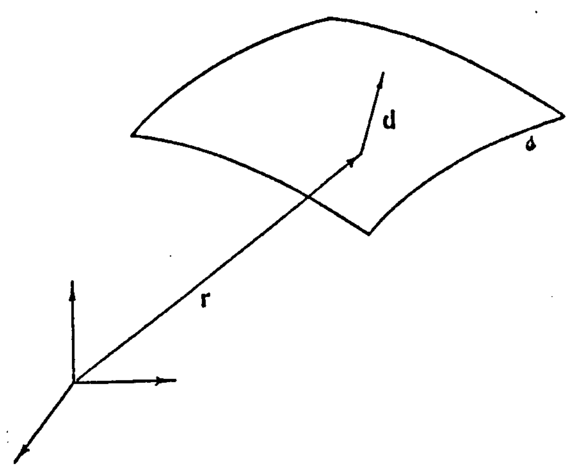

Figure 6-3
Cosserat surface

Step 3. Continuum modeling of composite laminate

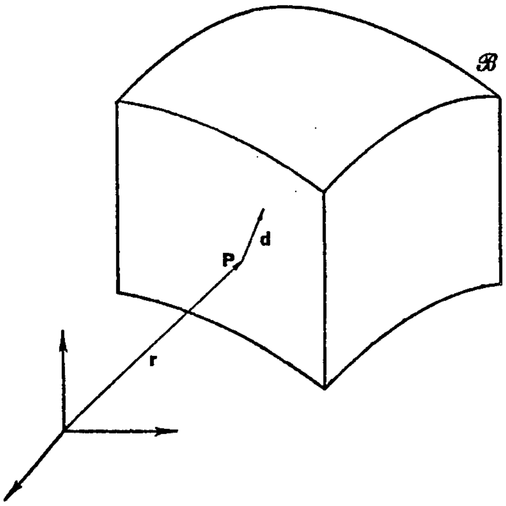

Figure 6-4
Macro-structure

It is to be emphasized that in the present discussion each Cosserat surface, i.e., micro-structure, is itself a three dimensional shell-like body $\mathscr{B}^*$ consisting of two layers of different homogeneous materials. We also notice that the material points within each representative element $\mathscr{B}^*$ are regular particles in the sense of classical continuum mechanics while the material points of the macro-structure are endowed not only with an assigned mass density but also with a director. We will refer to the body $\mathscr{B}$ as composite laminate, macro-continuum or macro-structure and to the body $\mathscr{B}^*$ as representative element, micro-continuum or micro-structure. Also, we will refer to particles of $\mathscr{B}$ as macro-particles or composite particles while the particles of the micro-structures will be referred to as micro-particles or simply particles (material points).

7. Coordinate systems for a composite laminate

At each point P of the macro-body $\mathscr{B}$ we introduce two sets of convected coordinates. One set of coordinates, denoted by $\eta^i = \{\eta^\alpha, \xi\}$, is used to describe the behavior of the micro-structure and the second set of coordinates, denoted by θ^i $(i = 1,2,3)$, is used to describe the behavior of the macro-structure. We assume that transformation from θ^i to η^i exists and possesses a unique inverse, i.e.,

$$\theta^i = \theta^i(\eta^k) = \theta^i(\eta^1, \eta^2, \eta^3) \tag{7.1}$$

and

$$\det\left(\frac{\partial \theta^i}{\partial \eta^j}\right) \neq 0 \tag{7.2}$$

We also make the additional assumption that

$$\theta^\alpha = \eta^\alpha \quad (\alpha = 1,2) \tag{7.3}$$

$$\theta^3 = \frac{1}{\varepsilon}\eta^3 = \frac{1}{\varepsilon}\xi \quad , \quad \varepsilon \ll 1 \tag{7.4}$$

The first of the above assumptions is for convenience while the second is required to account for change of scale from micro to macro-structure.

8. Definition of micro-structure

Within the context of three-dimensional classical continuum mechanics, consider a body $\mathscr{B}^*$ in the present configuration and let its boundary be a closed surface, denoted by $\partial\mathscr{B}^*$, and be composed of the following material surfaces:

a) The material surfaces

$$s_0 : \xi = 0 \qquad\qquad \xi_2 > 0 \qquad\qquad s_2 : \xi = \xi_2(\eta^\alpha) \tag{8.1}$$

b) The material surface

$$s_\ell : f(\eta^\alpha) = 0 \tag{8.2}$$

such that ξ = const. are closed smooth curves on the surface (8.2). We also consider a material surface of the form

$$s_1 : \xi = \xi_1(\eta^\alpha) \quad 0 < \xi_1 < \xi_2 \tag{8.3}$$

lying entirely between s_0 and s_2. We will refer to the surfaces defined above as follows.

a) s_0 : bottom face of the micro-structure

b) s_1 : interface of the micro-structure

c) s_2 : top face of the micro-structure

d) s_ℓ : lateral surface or normal surface of the micro-structure.

Let the part of the body $\mathscr{B}^*$ which is enclosed by the surfaces s_0, s_1 and s_ℓ be designated by $\mathscr{B}_1^*$ and the part enclosed by the surfaces s_1, s_2 and s_ℓ be denoted by $\mathscr{B}_2^*$. The field quantities associated by the parts $\mathscr{B}_1^*$ and $\mathscr{B}_2^*$ will be designated by subscripts 1 and 2 when necessary.

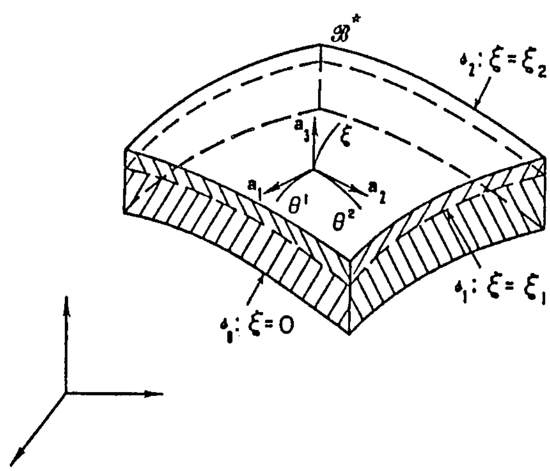

Figure 8-1
A typical micro-structure

9. Kinematics of micro- and macro-structures

We begin our development of the kinematical results by assuming that the position vector of a particle P^* of a representative element (micro-structure), i.e., $p^*(\eta^\alpha,\xi,\theta^3,t)$ in the present configuration has the form

$$p^* = r(\eta^\alpha,\theta^3,t) + \xi(\theta^3)d(\eta^\alpha,\theta^3,t) \tag{9.1}$$

The dual of (9.1) in a reference configuration is given by

$$P^* = R(\eta^\alpha,\theta^3) + \xi(\theta^3)D(\eta^\alpha,\theta^3) \tag{9.2}$$

If the reference configuration is taken to be the initial configuration at time $t = 0$, we obtain

$$\begin{aligned} p^*(\eta^\alpha,\xi,\theta^3,0) &= r(\eta^\alpha,\theta^3,0) + \xi d(\eta^\alpha,\theta^3,0) \\ &= R(\eta^\alpha,\theta^3) + \xi D(\eta^\alpha,\theta^3) \\ &= P^*(\eta^\alpha,\xi,\theta^3) \end{aligned} \tag{9.3}$$

The velocity vector v^* of the micro-structure at time t is given by

$$v^* = \frac{\partial p^*(\eta^\alpha,\xi,\theta^3,t)}{\partial t} = \dot{p}^*(\eta^\alpha,\xi,\theta^3,t) \tag{9.4}$$

where a superposed dot denotes the material time derivative, holding η^i and θ^i fixed. From (9.1) and (9.4) we obtain

$$v^* = v + \xi w \tag{9.5}$$

where

$$v = \dot{r} \quad , \quad w = \dot{d} \tag{9.6}$$

are the velocity and the director velocity of the macro-structure, respectively. From (9.1) we have

$$g^*_\alpha = a_\alpha + \xi \frac{\partial d}{\partial \eta^\alpha} \quad , \quad g^*_3 = d \tag{9.7}$$

where a_α are the surface base vectors of the surface s_0.

10. Basic field equations for the micro-structure

Making use of the theory of Cosserat (directed) surfaces after appropriate integration of the classical three-dimensional equations of motion, across the thickness of the micro-structure, we obtain the basic field equations for the shell-like micro-structure as follows:

$$\begin{aligned} &a: \ \overline{(\hat{\rho} a^{1/2})} = 0 \\ &b: \ \hat{\rho} a^{1/2}(\dot{v} + y^1\dot{w}) = (N^\alpha a^{1/2})_{,\alpha} + \hat{\rho}\hat{f}a^{1/2} \\ &c: \ \hat{\rho} a^{1/2}(y^1\dot{v} + y^2\dot{w}) = (M^\alpha a^{1/2})_{,\alpha} - m a^{1/2} + \hat{\rho}\hat{l}a^{1/2} \\ &d: \ a_\alpha \times N^\alpha + d \times m + d_{,\alpha} \times M^\alpha = 0 \\ &e: \ \hat{\rho}(\dot{\hat{\varepsilon}}) = N^\alpha \cdot v_{,\alpha} + M^\alpha \cdot w_{,\alpha} + m \cdot w \end{aligned} \tag{10.1}$$

where the various field quantities appear in (10.1) are

$\hat{\rho} = \hat{\rho}(\theta^\alpha,t)$:	The mass density of the micro-structure in the present configuration
$\nu = \nu(\theta^\alpha,t)$:	The outward unit normal to the boundary $\partial\hat{\mathcal{P}}$ of the micro-structure
$N^\alpha = N^\alpha(\theta^\alpha,t;\nu)$:	The resultant force per unit length of a curve in the present configuration
$M^\alpha = M^\alpha(\theta^\alpha,t;\nu)$:	The resultant couple per unit length of a curve in the present configuration
$\hat{f} = \hat{f}(\theta^\alpha,t)$:	The assigned force per unit mass of the micro-structure
$\hat{l} = \hat{l}(\theta^\alpha,t)$:	The assigned director force per unit mass of the micro-structure
$m = m(\theta^\alpha,t)$:	The intrinsic director force per unit area of the micro-structure
$y^\alpha = y^\alpha(\theta^\alpha)$:	The inertia coefficients
$\hat{\varepsilon} = \hat{\varepsilon}(\theta^\alpha,t)$:	The specific internal energy per unit mass of the micro-structure

The relations between the above field quantities and the field quantities in classical three-dimensional continuum mechanics are given below:

$$\hat{\rho} a^{1/2} = \int_0^{\xi_2} \rho^* g^{*1/2} d\xi = \int_0^{\xi_1} \rho_1^* g^{*1/2} d\xi + \int_{\xi_1}^{\xi_2} \rho_2^* g^{*1/2} d\xi \tag{10.2}$$

$$\hat{\rho} a^{1/2} y^\alpha = \int_0^{\xi_2} \rho^* g^{*1/2} \xi^\alpha d\xi = \int_0^{\xi_1} \rho_1^* g^{*1/2} \xi^\alpha d\xi + \int_{\xi_1}^{\xi_2} \rho_2^* g^{*1/2} \xi^\alpha d\xi \tag{10.3}$$

$$N^\alpha a^{1/2} = \int_0^{\xi_2} T^{*\alpha} d\xi = \int_0^{\xi_1} T^{*\alpha} d\xi + \int_{\xi_1}^{\xi_2} T^{*\alpha} d\xi \tag{10.4}$$

$$M^\alpha a^{1/2} = \int_0^{\xi_2} T^{*\alpha}\xi d\xi = \int_0^{\xi_1} T^{*\alpha}\xi d\xi + \int_{\xi_1}^{\xi_2} T^{*\alpha}\xi d\xi \tag{10.5}$$

$$m a^{1/2} = \int_0^{\xi_2} T^{*3} d\xi = \int_0^{\xi_1} T^{*3} d\xi + \int_{\xi_1}^{\xi_2} T^{*3} d\xi \tag{10.6}$$

$$\hat{\rho}\hat{f} a^{1/2} = \int_0^{\xi_2} \rho^* b^* g^{*1/2} d\xi + [T^{*3}]_{\xi=0}^{\xi=\xi_2} \tag{10.7}$$

$$\hat{\rho}\hat{l} a^{1/2} = \int_0^{\xi_2} \rho^* b^* g^{*1/2} \xi d\xi + [T^{*3}\xi]_{\xi=0}^{\xi=\xi_2} \tag{10.8}$$

11. Conservation laws for the micro-structure

The conservation laws for the micro-structure $\hat{\mathcal{P}}$, bounded by $\partial\hat{\mathcal{P}}$, may be obtained by integration of $(10.1)_a$ to $(10.1)_e$ over appropriate range of integration of the micro-structure (i.e., Cosserat surface). In this fashion we obtain

$$\begin{aligned} &a: \ \frac{d}{dt}\int_{\hat{\mathcal{P}}} \hat{\rho}\, d\sigma = 0 \\ &b: \ \frac{d}{dt}\int_{\hat{\mathcal{P}}} \hat{\rho}(v + y^1 w)\, d\sigma = \int_{\hat{\mathcal{P}}} \hat{\rho}\hat{f}\, d\sigma + \int_{\partial\hat{\mathcal{P}}} N\, ds \\ &c: \ \frac{d}{dt}\int_{\hat{\mathcal{P}}} \hat{\rho}(y^1 v + y^2 w)\, d\sigma = \int_{\hat{\mathcal{P}}} (\hat{\rho}\hat{l} - m)\, d\sigma + \int_{\partial\hat{\mathcal{P}}} M\, ds \\ &d: \ \frac{d}{dt}\int_{\hat{\mathcal{P}}} \hat{\rho}[r \times (v + y^1 w) + d \times (y^1 v + y^2 w)]\, d\sigma = \\ &\qquad \int_{\hat{\mathcal{P}}} \hat{\rho}(r \times \hat{f} + d \times \hat{l})\, d\sigma + \int_{\partial\hat{\mathcal{P}}} (r \times N + d \times M)\, ds \\ &e: \ \frac{d}{dt}\int_{\hat{\mathcal{P}}} \hat{\rho}(\hat{\varepsilon} + \mathcal{K})\, d\sigma = \int_{\hat{\mathcal{P}}} \hat{\rho}(\hat{f} \cdot v + \hat{l} \cdot w)\, d\sigma + \\ &\qquad \int_{\partial\hat{\mathcal{P}}} (N \cdot v + M \cdot w)\, ds \end{aligned} \tag{11.1}$$

where $\hat{\mathcal{K}}$ is the kinetic energy per unit mass of the micro-structure and $\hat{\mathcal{P}}$ is an arbitrary part of the Cosserat surface (i.e., micro-structure) with its boundary curve $\partial\hat{\mathcal{P}}$. The first of (11.1) is a mathematical statement of the conservation of mass, the second that of the linear momentum, the third is the conservation of the director momentum, the fourth that of the moment of momentum, and the fifth is the conservation of energy.

In (11.1) the micro-structure's contact force N and contact couple M (director force) are defined by

$$\int_{\hat{\mathcal{P}}} \mathrm{N}ds=\int_{\partial\mathcal{P}l} \mathrm{t}^*da \;\;,\;\; \int_{\hat{\mathcal{P}}} \mathrm{M}ds=\int_{\partial\mathcal{P}l} \mathrm{t}^*\xi d\xi \qquad (11.2)$$

and are related to N^α and M^α as follows:

$$\mathrm{N}=\mathrm{N}^\alpha v_\alpha \;\;,\;\; \mathrm{M}=\mathrm{M}^\alpha v_\alpha \qquad (11.3)$$

where v_α ($\alpha = 1,2$) are covariant components of v.

12. Conservation laws for composite laminates

We recall that the composite laminate is assumed to consist of infinitely many Cosserat surfaces. This assumption is justified by physical considerations since the thickness of each ply is small in comparison with the thickenss of the laminate itself. Conservation laws for composite laminates (i.e., macro-structure) may now be obtained by integrating $(11.1)_a$ to $(11.1)_e$ with respect to θ^3 and over a relevant range of integration. In this manner and with reference to the present configuration, we obtain

$$\mathrm{a}\;:\; \frac{d}{dt}\int_{\mathcal{P}} \rho dv=0$$

$$\mathrm{b}\;:\; \frac{d}{dt}\int_{\mathcal{P}} \rho(\mathrm{v}+y^1\mathrm{w})dv=\int_{\mathcal{P}} \rho \mathrm{b} dv+\int_{\partial\mathcal{P}} \mathrm{t} da$$

$$\mathrm{c}\;:\; \frac{d}{dt}\int_{\mathcal{P}} \rho(y^1\mathrm{v}+y^2\mathrm{w})dv=\int_{\mathcal{P}} (\rho\mathrm{c}-\mathrm{k})dv+\int_{\partial\mathcal{P}} \mathrm{s} da$$

$$\mathrm{d}\;:\; \frac{d}{dt}\int_{\mathcal{P}} \{\mathrm{r}\times(\mathrm{v}+y^1\mathrm{w})+\mathrm{d}\times(y^1\mathrm{v}+y^2\mathrm{w})\}dv= \qquad (12.1)$$

$$\int_{\mathcal{P}} \rho(\mathrm{r}\times\mathrm{b}+\mathrm{d}\times\mathrm{c})dv+\int_{\partial\mathcal{P}} (\mathrm{r}\times\mathrm{t}+\mathrm{d}\times\mathrm{s})da$$

$$\mathrm{e}\;:\; \frac{d}{dt}\int_{\mathcal{P}} \rho(\varepsilon+\mathcal{K})dv=\int_{\mathcal{P}} \rho(\mathrm{b}\cdot\mathrm{v}+\mathrm{c}\cdot\mathrm{w})dv+$$

$$\int_{\partial\mathcal{P}} (\mathrm{t}\cdot\mathrm{v}+\mathrm{s}\cdot\mathrm{w})da$$

The first of (12.1) is the mathematical statement of conservation of mass, the second that of linear momentum principle, the third that of director momentum, the fourth is the principle of moment of momentum, and the fifth represents the balance of energy for composite laminates.

We observe that the basic structures of $(12.1)_{a,b}$ and their forms are analogous to the corresponding conservation laws of the classical 3-dimensional continuum mechanics. Equation $(12.1)_c$ does not exist in the classical continuum mechanics while equations $(12.1)_{d,e}$ although exist they have simpler forms.

In (12.1) r and d denote the *position vector* and the *director* associated with a *composite particle*, respectively, while the velocity and the director velocity of the composite particle are given by v and w. The definition of the various field quantities in (12.1) and their relation to their counterparts in micro-structure and the similar three dimensional quantities are given below.

1) $\rho=\rho(\theta^i,t)$ is the *composite assigned mass density* in the present configuration given by

$$\rho g^{1/2}=\hat{\rho} a^{1/2}=\int_0^{\xi_2} \rho^* g^{*1/2} d\xi \qquad (12.2)$$

where in (12.2) $\hat{\rho}$ is the mass density of the micro-structure, ρ^* is the classical 3-dimensional mass density, g is the determinant of the metric tensor g_{ij} associated with the composite coordinate system θ^i, g^* is the determinant of the metric tensor g^*_{ij} associated with the micro-structure coordinate system $\eta^i=\{\eta^\alpha,\xi\}=\{\theta^\alpha,\xi\}$, a is the determinant of the two-dimensional (surface) metric tensor $a_{\alpha\beta}$ associated with the Cosserat surface (micro-structure).

We notice that the dimensions of ρ^* and $\hat{\rho}$ are mass per unit volume and mass per unit area, respectively. However, the dimension of ρ is the dimension of the weighted integral of the mass.

2) $\mathrm{b}=\mathrm{b}(\theta^i,t)$ is the *composite assigned body force density* per unit of ρ, given by

$$\rho g^{1/2}\mathrm{b}=\int_0^{\xi_2} \rho^* g^{*1/2}\mathrm{b}^* d\xi \qquad (12.4)$$

where b^* is the classical 3-dimensional body force density. The dimension of b should be clear from (12.4).

3) $\mathrm{c}=\mathrm{c}(\theta^i,t)$ is the *composite assigned body couple density* per unit of ρ, given by

$$\rho g^{1/2}\mathrm{c}=\int_0^{\xi_2} \rho^* g^{*1/2}\mathrm{b}^*\xi d\xi \qquad (12.5)$$

The dimension of c should be clear from (12.5).

4) $\mathrm{t}=\mathrm{t}(\theta^i,t;\mathrm{n})$ is the *composite assigned contact force (stress vector)* (per unit area of the composite) such that

$$\mathrm{t}=g^{-1/2}\mathrm{T}^i n_i \qquad (12.6)$$

$$\mathrm{T}^i{}_{,i}=\int_0^{\xi_2} \mathrm{T}^{*i}{}_{,i} d\xi \qquad (12.7)$$

$$\mathrm{T}^\alpha=\int_0^{\xi_2} \mathrm{T}^{*\alpha} d\xi=a^{1/2}\mathrm{N}^\alpha \qquad (12.8)$$

$$\mathrm{T}^3{}_{,3}=\mathrm{T}^{*3}|_{\xi=\xi_2}-\mathrm{T}^{*3}|_{\xi=0}=\Delta\,\mathrm{T}^{*3} \qquad (12.9)$$

where $\mathrm{n}=n^i g_i$ is the outward unit normal to the surface on which t acts, T^{*i} is the classical stress vector and N^α is the resultant force of the micro-structure (i.e., Cosserat surface). We also recall that a comma on the left-hand side of (12.7) and (12.9) denotes partial differentiation with respect to θ^i. However, a comma on the right-hand side of (12.7) denotes partial differentiation with respect to $\eta^i=\{\eta^\alpha,\xi\}$.

5) $\mathrm{s}=\mathrm{s}(\theta^i,t;\mathrm{n})$ is the *composite assigned contact couple (couple stress vector)* per unit area of the composite such that

$$\mathrm{s}=g^{-1/2}\mathrm{S}^i n_i \qquad (12.10)$$

$$\mathrm{S}^i{}_{,i}=\int_0^{\xi_2} \mathrm{T}^{*i}{}_{,i}\xi d\xi \qquad (12.11)$$

$$\mathrm{S}^\alpha=\int_0^{\xi} \mathrm{T}^{*\alpha}\xi d\xi=a^{1/2}\mathrm{M}^\alpha \qquad (12.12)$$

$$\mathrm{S}^3{}_{,3}=(\mathrm{T}^{*3}\xi)|_{\xi=\xi_2}-(\mathrm{T}^{*3}\xi)|_{\xi=0}=\Delta(\mathrm{T}^{*3}\xi) \qquad (12.13)$$

where M^α is the resultant couple of the micro-structure and the same remark as in (4) above holds for commas and partial differentiation in (12.11) and (12.13).

6) $\mathrm{k}=\mathrm{k}(\theta^i,t)$ is the *composite assigned intrinsic (director) force*, per unit volume of the composite, given by

$$g^{1/2}\mathrm{k}=a^{1/2}\mathrm{m}=\int_0^{\xi_2} \mathrm{T}^{*3} d\xi \qquad (12.14)$$

where m is the intrinsic director force of the micro-structure.

7) $y^\alpha=y^\alpha(\theta^i)$ are the inertia coefficients which are independent of time and are given by

$$y^\alpha = \int_0^{\xi_2} \rho^* g^{*1/2} \xi^\alpha d\xi \qquad (12.15)$$

8) $\varepsilon = \varepsilon(\theta^i, t)$ is the *composite assigned specific internal* energy per unit of ρ given by

$$\rho g^{1/2} \varepsilon = \hat{\rho} a^{1/2} \hat{\varepsilon} = \int_0^{\xi_2} \rho^* g^{*1/2} \varepsilon^* d\xi \qquad (12.16)$$

where ε^* is the classical 3-dimensional specific internal energy and $\hat{\varepsilon}$ is the specific internal energy per unit $\hat{\rho}$ for the micro-structure.

9) $\mathcal{K} = \mathcal{K}(\theta^i, t)$ is the *composite assigned specific kinetic energy* per unit of ρ and is given by

$$\mathcal{K} = \hat{\mathcal{K}} = \frac{1}{2} (\mathbf{v} \cdot \dot{\mathbf{v}} + 2y^1 \mathbf{v} \cdot \mathbf{w} + y^2 \mathbf{w} \cdot \mathbf{w}) \qquad (12.17)$$

where $\hat{\mathcal{K}}$ represents the kinetic energy per unit $\hat{\rho}$ of the micro-structure. The momentum corresponding to the velocity $\mathbf{v}$ and the director momentum corresponding to $\mathbf{w}$ are given by

$$\rho \frac{\partial \mathcal{K}}{\partial \mathbf{v}} = \rho(\mathbf{v} + y^1 \mathbf{w}) \qquad (12.18)$$

$$\rho \frac{\partial \mathcal{K}}{\partial \mathbf{w}} = \rho(y^1 \mathbf{v} + y^2 \mathbf{w}) \qquad (12.19)$$

For simplicity, when there is no possibility of confusion, we may adopt the following simplified terminology:

ρ: "composite mass density"
b: "composite body force density"
c: "composite body couple density"
t: "composite contact force (stress vector)"
s: "composite contact couple (couple stress vector)"
k: "composite intrinsic force"
ε: "composite specific internal energy"
$\mathcal{K}$: "composite specific kinetic energy"

13. Remarks on composite stress vector and composite couple stress vector

An important characteristic of the present theory is the introduction of the composite contact force, **t**, and the composite contact couple, **s**. It can be shown [4] that **t** and **s** have the properties

$$\mathbf{t}(\theta^i, t\,;\, \mathbf{n}) = -\,\mathbf{t}(\theta^i, t\,;\, -\mathbf{n}) \qquad (13.1)$$

and

$$\mathbf{s}(\theta^i, t\,;\, \mathbf{n}) = -\,\mathbf{s}(\theta^i, t\,;\, -\mathbf{n}) \qquad (13.2)$$

where **n** is the outward unit normal to a surface within the composite. According to the results (13.1) and (13.2), the composite stress vector and the composite couple stress vector acting on opposite sides of the same surface at a given point within the composite laminates are equal in magnitude and opposite in direction. In addition, it can be demonstrated [4] that $\mathbf{T}^i$ and $\mathbf{S}^i$ are expressible as

$$\mathbf{T}^i = g^{1/2} \tau^{ij} \mathbf{g}_j \qquad (13.3)$$

and

$$\mathbf{S}^i = g^{1/2} s^{ij} \mathbf{g}_j \qquad (13.4)$$

where τ^{ij} and s^{ij} are contravariant components of the *composite stress tensor* and the *composite couple stress tensor*. It can also be shown that $\mathbf{T}^3$ represent the interlaminar stress vector which is related to the three components of interlaminar stresses τ^{3j} ($j = 1,2,3$) through (13.3). While τ^{3j} represent the three components of stress tensor, the same is not true for the interlaminar couple stress tensors. In fact it can be proven that $\mathbf{S}^3$ and s^{3j} ($j = 1,2,3$) vanish identically, i.e.,

$$\mathbf{S}^3 = 0 \ , \ s^{3j} = 0 \qquad (13.5)$$

We notice that the composite stress, the composite couple stress vector and the composite intrinsic force are to be specified by constitutive relations. Hence, there exist eighteen consitutive relations in the present theory.

14. Basic field equations for composite laminates

The basic field equations for composite laminates follow from $(12.1)_a$ to $(12.1)_e$ and are given by

$$
\begin{aligned}
a &: \ \dot{\rho} + \frac{\dot{g}}{2g}\rho = 0 \\
b &: \ \mathbf{T}^i{}_{,i} + \rho g^{1/2}\mathbf{b} = \rho g^{1/2}(\dot{\mathbf{v}} + y^1\dot{\mathbf{w}}) \\
c &: \ \mathbf{S}^i{}_{,i} + g^{1/2}(\rho\mathbf{c} - \mathbf{k}) = \rho g^{1/2}(y^1\dot{\mathbf{v}} + y^2\dot{\mathbf{w}}) \qquad (14.1) \\
d &: \ \mathbf{g}_i \times \mathbf{T}^i + \mathbf{d}_{,i} \times \mathbf{S}^i + g^{1/2}\mathbf{d} \times \mathbf{k} = 0 \\
e &: \ \rho g^{1/2}\dot{\varepsilon} = \mathbf{T}^i \cdot \mathbf{v}_{,i} + \mathbf{S}^i \cdot \mathbf{w}_{,i} + g^{1/2}\mathbf{k} \cdot \mathbf{w} = g^{1/2}P
\end{aligned}
$$

where P is the mechanical power per element of volume of the composite and is given by

$$g^{1/2}P = \mathbf{T}^i \cdot \mathbf{v}_{,i} + \mathbf{S}^i \cdot \mathbf{w}_{,i} + g^{1/2}\mathbf{k} \cdot \mathbf{w} \qquad (14.2)$$

The basic field equations (14.1) when expressed in component forms will reduce to

$$
\begin{aligned}
a &: \ \dot{\rho} + \frac{\dot{g}}{2g} = 0 \\
b &: \ \tau^{ij}{}_{|i} + \rho b^j = \rho(\alpha^j + y^1\beta^j) \\
c &: \ s^{ij}{}_{|i} + (\rho c^j - k^j) = \rho(y^1\alpha^j + y^2\beta^j) \qquad (14.3) \\
d &: \ \varepsilon_{ijn}(\tau^{ij} + d^i{}_{|m}s^{mj} + d^i k^j) = 0 \\
e &: \ \rho\,\dot{\varepsilon} = \tau^{ij}v_{j|i} + s^{ij}w_{j|i} + k^i w_i = P
\end{aligned}
$$

while the expression for mechanical power takes the form

$$P = \tau^{ij}v_{j|i} + s^{ij}w_{j|i} + k^i w_i \qquad (14.4)$$

In $(14.3)_{b,c}$, α^j and β^j are the contravariant components of $\dot{\mathbf{v}}$ and $\dot{\mathbf{w}}$. With reference to equation $(14.3)_d$ we observe that the symmetry of the stress tensor is not valid. It may be shown that in the absence of the micro-structure and the director the basic field equations (14.1) or (14.3) as well as the expressions for power reduce to those of classical continuum mechanics.

Remarks and future work

In addition to conservation laws, a set of eighteen nonlinear constitutive relations has been developed for elastic composite laminates [4] which are similar, in form, to those of classical continuum mechanics. Also, appropriate equations for kinematical constraints have been obtained and consequently a constraint theory of composite laminates has been developed. The linearization of the theory has been carried out and the complete set of linearized equations has been specialized to the cases of initially flat and initially cylindrical composite laminates.

The future works include the application of the theory to static and dynamic problems. In particular, the effect of interlaminar stresses on behavior and integrity of composite laminates under both static and dynamic loading conditions shall be studied. In addition, the theory shall be further developed to include the effect of anisotropy, temperature and multi-layering. Finally, the variational formulation of the present theory and development of a numerical strategy for finite element discretization is to be carried out.

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AXISYMMETRIC FAILURE MODEL FOR BRITTLE MATRIX COMPOSITES

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The principal failure modes that take place in the tensile loading of 0° unidirectional brittle matrix composites (BMC), such as those having ceramic and glass-ceramic matrices, are matrix cracking, debonding (probably with friction), and fiber breaks. In order to predict the influence of these forms of damage, including their possible interactions, we shall develop a model which attempts to accurately represent the stress field in their presence. Our approach is to appeal to the Reissner variational theorem [1], which has been successfully employed to study the elastic stress fields in flat laminates [2] as well as involute bodies of revolution [3]. It has been demonstrated [2, 3] that such models provide a reasonable description for the stress field in the vicinity of a stress riser, even though no singularity is present, so that they are appropriate for use in conjunction with an average stress failure theory [4, 5] or energy release rate criterion. It has also been shown that these models incorporate the capability to provide fine sublayers to improve solution accuracy.

The medium considered here consists of a solid circular cylindrical body (fiber) which is surrounded by a number of concentric cylindrical annuli. Cylindrical coordinates (r, z) are employed. Each sublayer is represented by an index k (k = 0, 1, ---N) where k = 0 refers to the fiber and k = 1, ---N denote the annuli in the direction if increasing r. Damage is simulated by axisymmetric cracks which are disc-shaped in the fiber, annular planes in the media k = 1, 2---N, and/or cylindrical debonds of arbitrary constant height Δz at an arbitrary interface. We assume that the (axisymmetric) traction and/or displacement boundary conditions are known. In the application of the model, the various annuli may be used to represent coating or interphase regions, matrix, and/or regions which are assigned effective composite properties. Alternatively, other methods of approximate heterogeneous elasticity solutions might be used to establish the boundary conditions on a local region of the composite, which could then be represented by the present model.

In order to investigate the quality of the approximate model, the symmetric pullout problem treated by Kurtz [6] has been solved. Results are given for a uniform temperature change for the case of zero applied force on the fiber. The interfacial stresses σ_r and τ_{rz} are seen to agree very closely for the two solutions, while the stresses σ_θ and σ_z in the matrix at the interface are somewhat underestimated in the present model since they have been assumed to be linear functions of r. Subdividing the matrix region into 2 annuli could presumably bring the two solutions into even closer agreement. These preliminary results are quite encouraging for the application of approximate models of this type in the study of failure of BMC.

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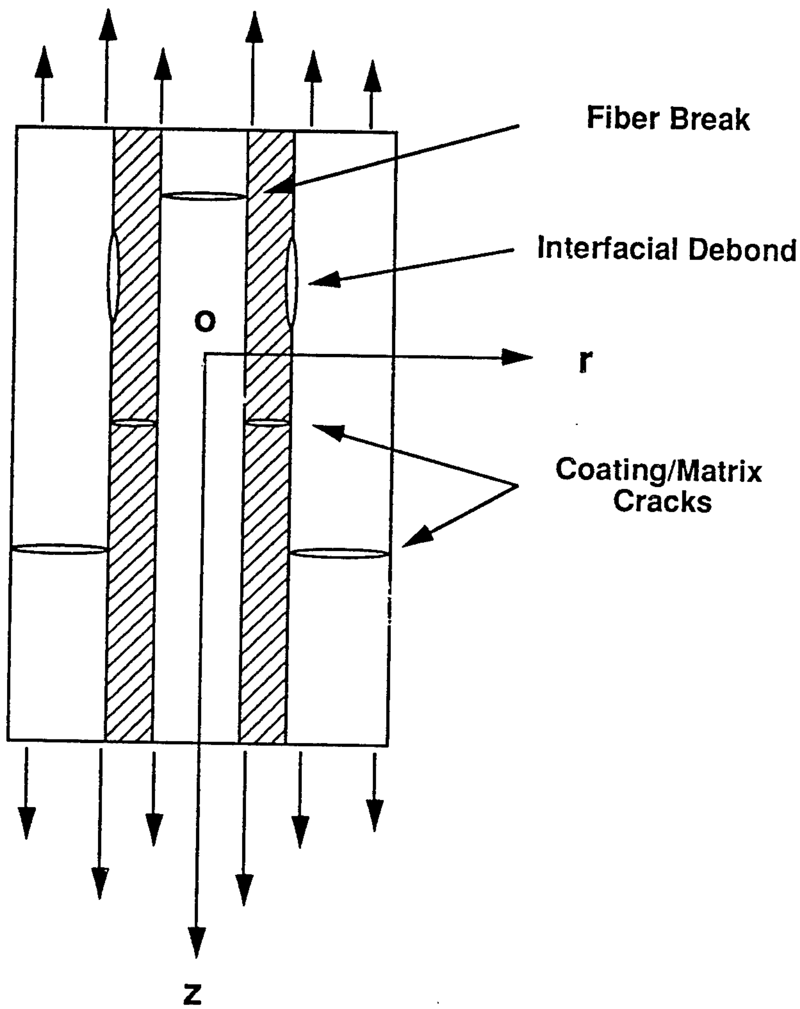
AXISYMMETRIC DAMAGE MODEL



REISSNER'S VARIATIONAL PRINCIPLE

Setting

$$\delta J = 0$$

where

$$J = \int_V F\,dv - \int_{S'} \tilde{T}_i U_i\,dS$$

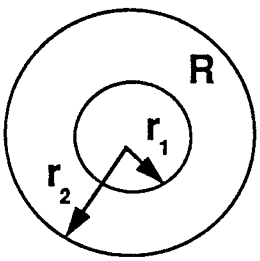
and

$$F = \frac{1}{2}\tau_{ij}(U_{i,j} + U_{j,i}) - W(\tau_{ij}, e_{ij})$$

W= complementary energy
S' = Traction-prescribed boundary

leads to the governing eqs. (and B. C.) of elasticity.

FORMULATION OF MODEL



In Each Region R, Assume

$$(\mathbf{r}_2-\mathbf{r}_1)\sigma_i = \mathbf{P}_{i1}(\mathbf{z})(\mathbf{r}_2-\mathbf{r})+\mathbf{P}_{i2}(\mathbf{z})(\mathbf{r}-\mathbf{r}_1) : \mathbf{i}=\mathbf{z},\theta$$

Then Compute

$$(\mathbf{r}\sigma_{\mathbf{rz}})_{,\mathbf{r}} = -\mathbf{r}\sigma_{\mathbf{z},\mathbf{z}}$$

$$(\mathbf{r}\sigma_{\mathbf{r}})_{,\mathbf{r}} = \sigma_\theta-\mathbf{r}\sigma_{\mathbf{rz},\mathbf{z}}$$

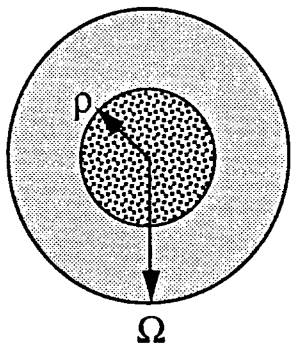
and

$$(\overline{\mathbf{q}}, \mathbf{q}^*, \widehat{\mathbf{q}}, \overline{\overline{\mathbf{q}}}) = \int_{r_1}^{r_2} \mathbf{q}\,(1, \mathbf{r}, \mathbf{r}^2, \mathbf{r}^3)\,\mathbf{dr} : \mathbf{q} = \mathbf{u}, \mathbf{w}$$

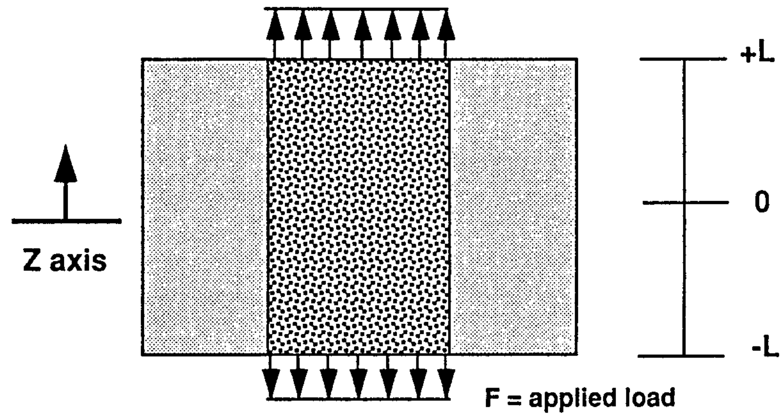
Substitute into

$$\delta\mathbf{J} = 0$$

For N Rings This Leads to
18N + 16 Algebraic & O.D.E. in Z with 5N + 4 B.C.



FIBER-MATRIX TOP VIEW



FIBER-MATRIX SIDE VIEW

STRESSES IN MATRIX AT INTERFACE

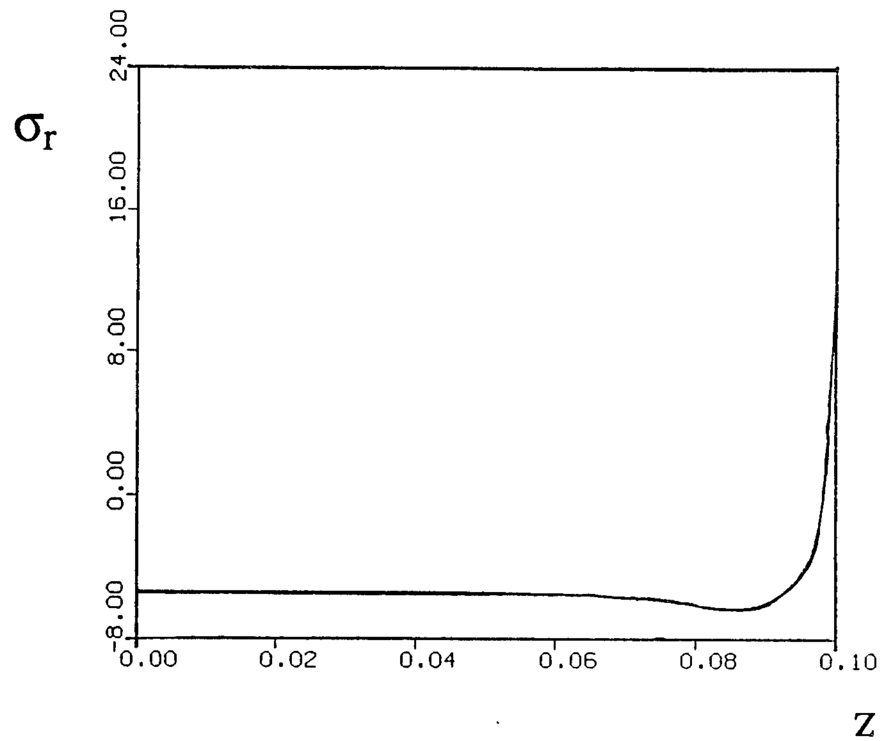
FOR

$\Delta T = -.5555^{\circ}C$

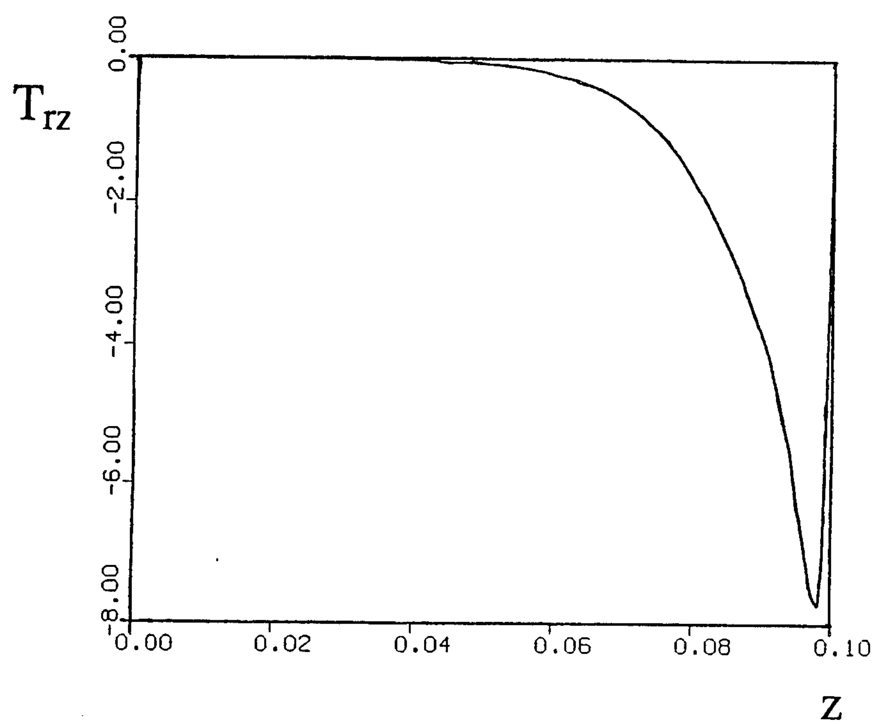
$E_f = 60Msi$
$\nu_f = .2$
$\alpha_f = 3.25 \times 10^{-6}/^{0}C$

$E_m = 9.1Msi$
$\nu_m = .2$
$\alpha_m = 3.5 \times 10^{-6}/^{0}C$

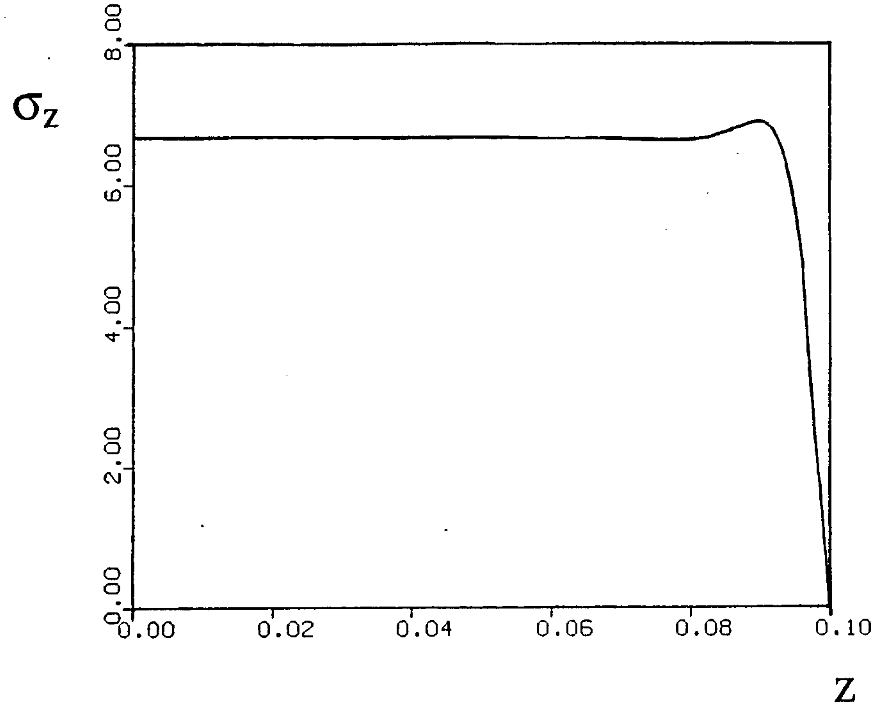
$\rho = .01"$, $\Omega = .02"$, $L = .1"$

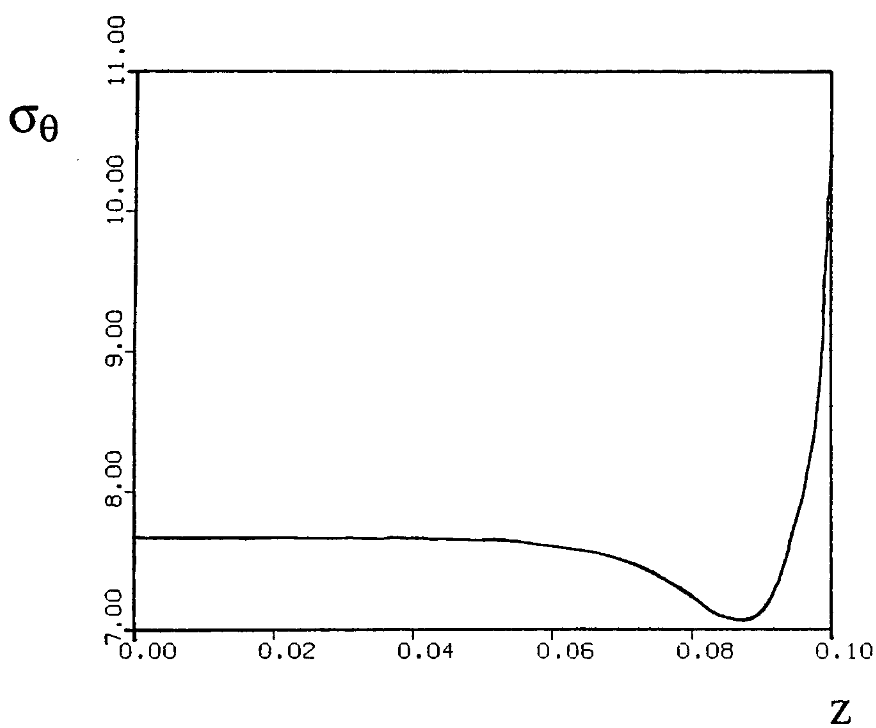


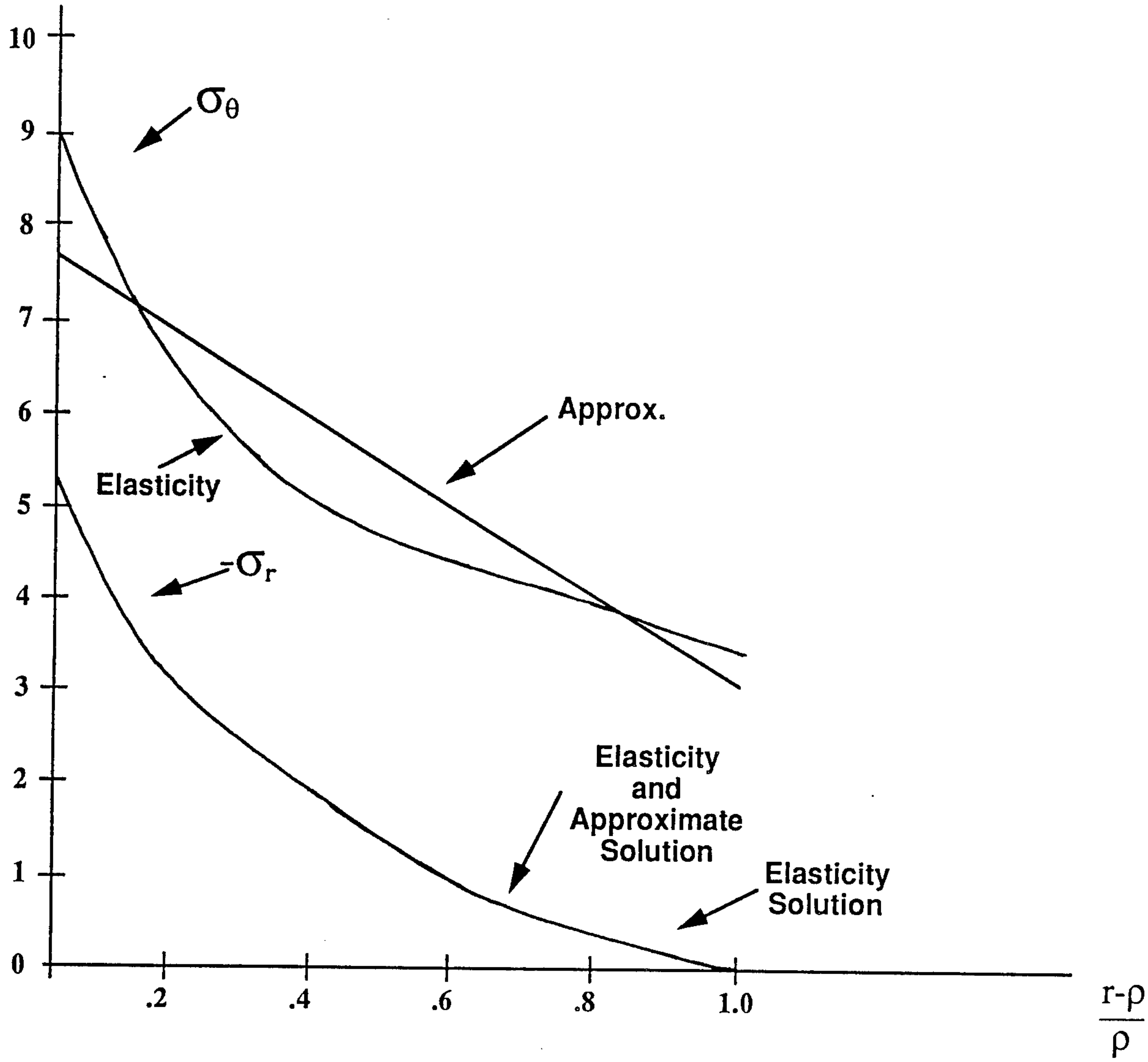
σ_r
24.00
16.00
8.00
0.00
-8.00
0.00
0.02
0.04
0.06
0.08
0.10
Z



T_{rz}
0.00
-2.00
-4.00
-6.00
-8.00
0.00
0.02
0.04
0.06
0.08
0.10
Z


σz
8.00
6.00
4.00
2.00
0.00
0.00
0.02
0.04
0.06
0.08
0.10
Z


σθ
11.00
10.00
9.00
8.00
7.00
0.00
0.02
0.04
0.06
0.08
0.10
Z


10
9
8
7
6
5
4
3
2
1
0
σ_θ
Approx.
Elasticity
$-\sigma_r$
Elasticity and Approximate Solution
Elasticity Solution
.2
.4
.6
.8
1.0
$\frac{r-\rho}{\rho}$

STRENGTH REDUCTION DUE TO MATRIX FLAWS IN FIBER-REINFORCED CERAMICS

Paul S. Steif
Department of Mechanical Engineering
Carnegie Mellon University

ABSTRACT

It is increasingly accepted that the fiber-matrix interface plays a significant role in the fracture resistance of fiber-reinforced composites. For ceramic-matrix composites, in particular, a relatively weak interface generally promotes composite toughness, although sometimes at the expense of other properties such as transverse strength and shear strength. Under longitudinal tension, cracks growing in the matrix are diverted to a weak interface, thereby sparing the fiber, at least until higher levels of load are achieved. The existing view of this phenomenon - referred to here as the relative toughness analysis - is that the crack diverts to the interface if the interfacial fracture toughness is sufficiently less than the fiber fracture toughness.

The present work takes alternative approach, by asking the following questions: Given that there are matrix cracks which impinge upon fibers, what is the level of load at which the impinged fibers will break? Is that load greater than or less than the load necessary to propagate the matrix crack around the fiber? And, how is this process influenced by microstructural material parameters such as interface strength? Now, say it is determined, via the relative toughness analysis, that the matrix crack penetrates the fiber rather than diverting to the interface. Then, with the same analysis, one can predict the load at which the fiber would fail. On the other hand, say the relative toughness analysis predicts that the crack is diverted to the interface. Surely the fiber will fail at some higher level of remote load, whether or not the matrix crack first manages to propagate around the fiber. Furthermore, the increment in remote load which is needed to cause fiber failure is critical to determining the composite strength. Yet, current methods of analysis, including relative toughness arguments, are incapable of resolving this question.

In other words, the relative toughness analysis distinguishes only between those interfaces that divert matrix cracks and those that do not. The far more important requirement (from a practical point of view) of distinguishing between different interfaces *all of which do allow diversion* cannot be met with the relative toughness analysis. As evidence of the necessity of making such a distinction, consider fracture surfaces of ceramic-matrix composites: there are very fibrous (broom-like) fracture surfaces, and there are surfaces with very short pullout lengths. Only occasionally will the surface appear to be almost perfectly flat, with virtually no pullout. While the variety of pullout lengths is a consequence of the random distribution of flaws in the fibers, it undoubtedly necessitates prior debonding of the interface as well. Therefore, it would clearly be valuable to have a model which distinguishes among the vast majority of cases in which matrix cracks cause some debonding of the interface prior to failure of the fibers. Such a model would involve an evaluation of stresses acting on the fiber *after* the matrix crack has been diverted to the interface.

To provide this evaluation, at least in a two-dimensional context, we have carried out an analysis of a single crack impinging upon an interface which offers frictional resistance to sliding after debonding. Furthermore, a unified fiber failure criterion is employed which can be applied whether or not debonding has occurred. This analysis provides a means of determining the level of applied stress at which an impinged fiber will fail.

STRENGTH REDUCTION DUE TO MATRIX FLAWS
IN FIBER-REINFORCED CERAMICS

PAUL S. STEIF

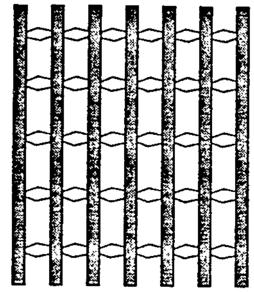
DEPARTMENT OF MECHANICAL ENGINEERING

CARNEGIE MELLON UNIVERSITY

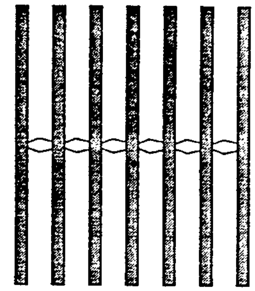
COLLABORATOR: ANNA DOLLAR

SUPPORT: AFOSR

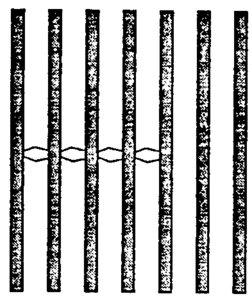
Various degrees of matrix cracking



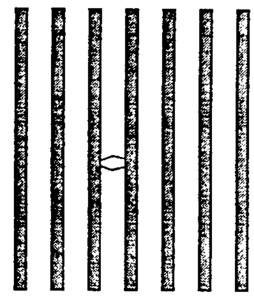
Multiple matrix cracking



Single matrix crack



Propagating matrix crack



Nascent matrix crack

For each of the above configurations,

one can ask:

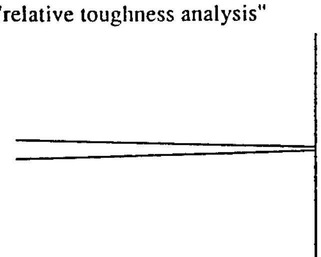
When will the impinged fibers fail?

Do the matrix cracks hasten fiber failure?

What role does the interface play?

One existing theory:

"relative toughness analysis"



What happens next?

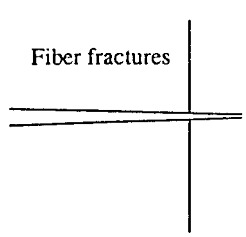
Composite's lore says:

A <u>strong</u> interfacial bond makes a <u>bad</u> ceramic-matrix composite-

matrix crack runs right through the fibers

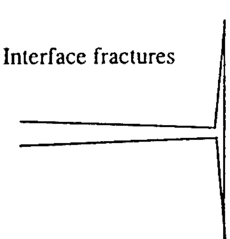
Quantitative theory is lacking

Fiber fractures



or

Interface fractures



Depends on relative toughness of fiber and interface

ABOVE THEORY IS <u>INSUFFICIENT</u>:

Distinguishes only between interfaces that allow fibers to break

and those that do not

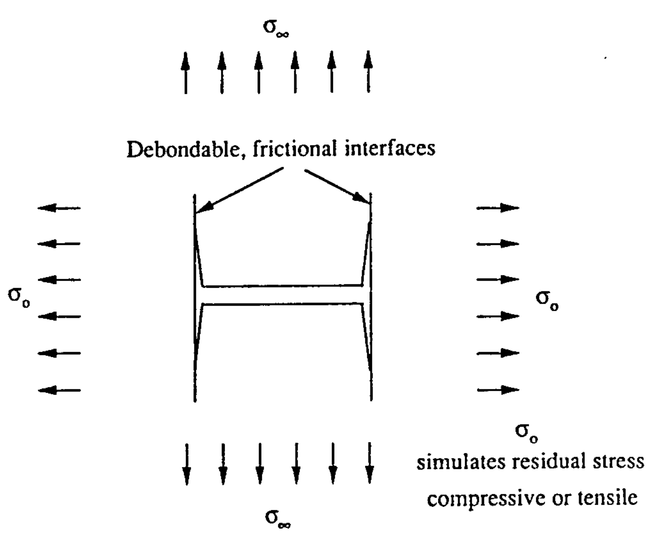
But in most composites, there is some debonding

before fiber failure

Must have a theory which distinguishes between interfaces, all of which allow some debonding

Theory for fiber failure in the presence of some debonding

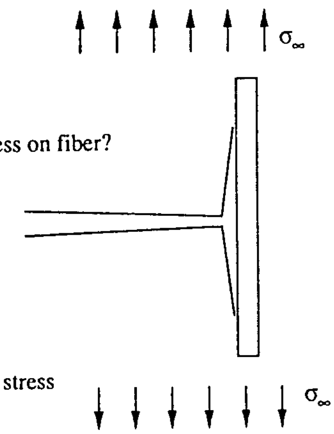
Following specific problem captures essential features



What is stress concentration ahead of crack

averaged over, say, a fiber diameter?

Under remote tension, say that interface debonds



Then, what is resultant stress on fiber?

Stress is <u>not</u> singular,

but it is <u>more</u> than the applied stress

Goal: Compute stress concentration as a function of:

1. Energy to debond interface
2. Friction at debonded interface
3. Applied stress

Specific analysis depends on degree of matrix cracking

Solution Method:

Debonding and slippage at interface is represented by a continuous distribution of dislocations

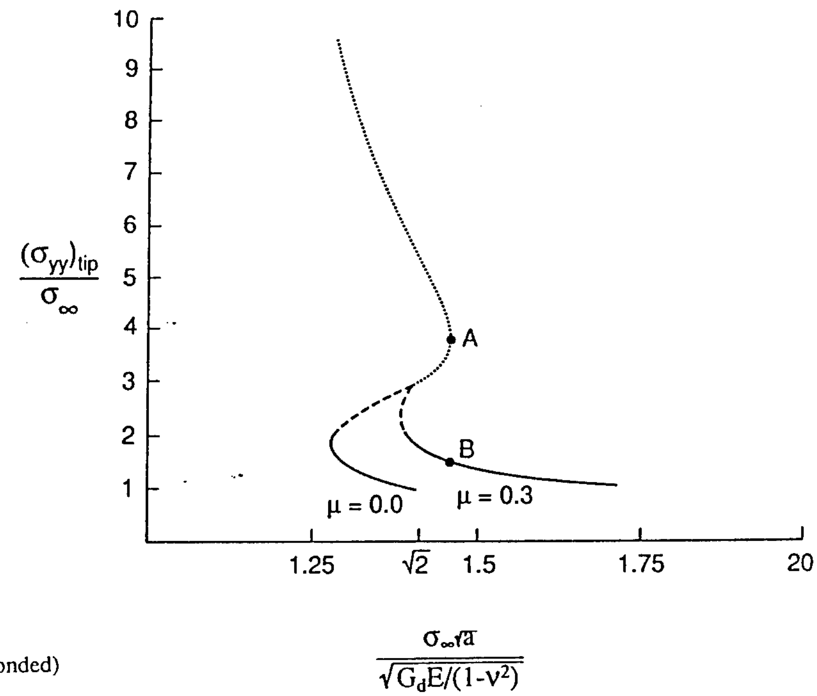
Reduces problem to a few integral equations

Integral equations are solved numerically

Final accuracy of solution is within 1 or 2 %.

Input Parameters:

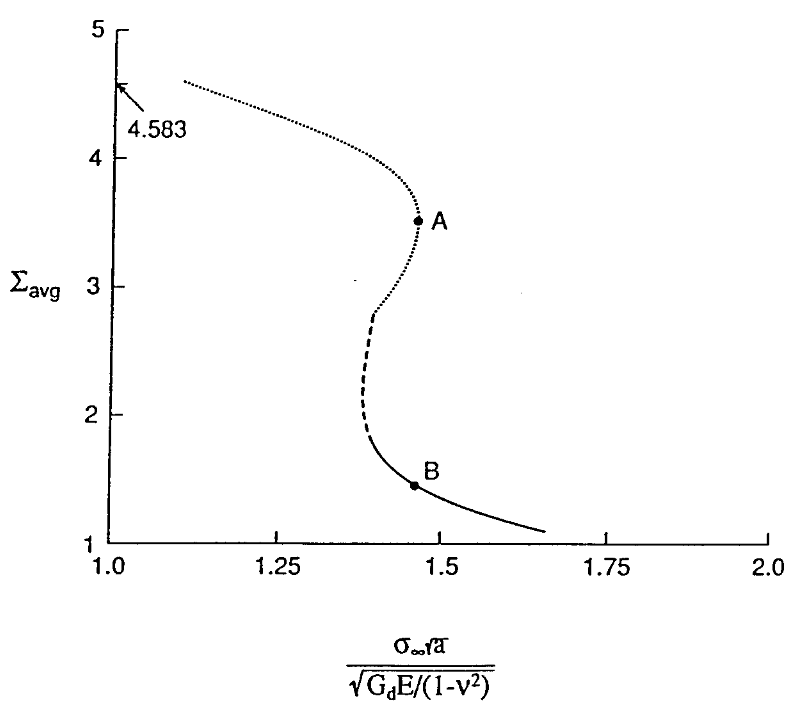
1. Crack length
2. Applied stress
3. Energy to debond
4. Residual stress at interface
5. Friction coefficient

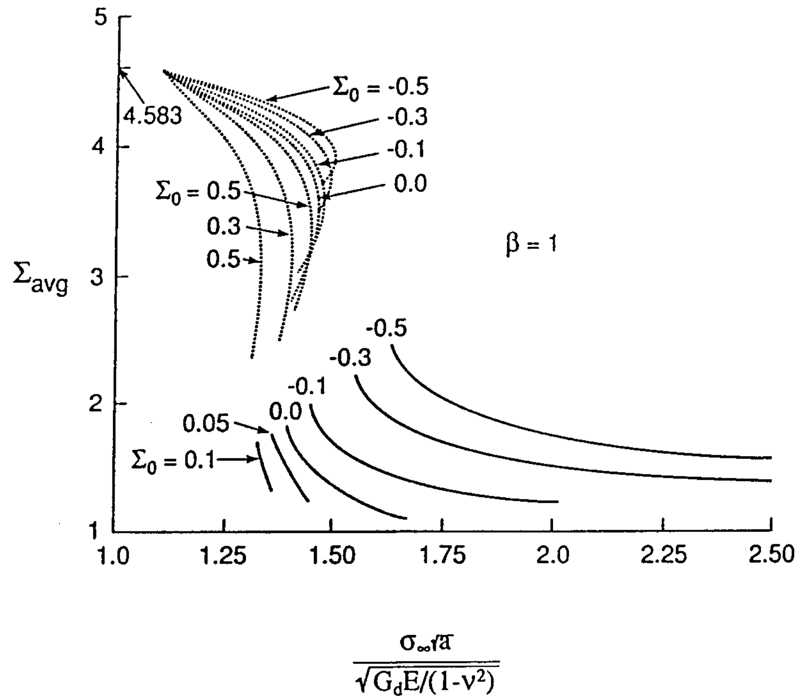


Quantities that can be obtained from solutions:

1. Extent of debonding
2. Stress at crack tip (it's finite, if interface has debonded)
3. Stress concentration averaged over fiber diameter

Average stress concentration goes into fiber failure criterion





CONCLUSIONS

1. Weaker bonding reduces tendency for fiber fracture

2. Lower interfacial friction reduces tendency for fiber fracture

3. Larger diameter fibers are better for flaw resistance

(Theory quantifies these improvements)

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MECHANICS OF COMPOSITES REVIEW

October 1990

Composite Materials Technology at the Ballistic Research Laboratory

Dr. Bruce P. Burns
Chief, Mechanics & Structures Branch
Ballistic Research Laboratory

BRL PROGRAM SCOPE

1. Technology to Sensibly Integrate Lightweight Materials into Projectiles

2. Advanced Processing Technology for Critical Projectile Components

3. Evaluation of Various Modular Armor Confinement Structures

4. Development of Vulnerability Assessment Methodology for Both Thin and Thick Section Structures

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LIGHTWEIGHT STRUCTURES

FOR

INTERIOR BALLISTICS

BALLISTIC RESEARCH LABORATORY

US ARMY LABORATORY COMMAND

SERVICE CONDITIONS

MAIN PRESSURE PULSE

Zero to 60,000 - 100,000 Kpsi
in
0.5 to 4 msec

ENGRAVING (SEALING) EFFECTS

Rotating Band/Obturator

Strains of 200%
Strain Rates of 2000/Sec

DESIGN PHILOSOPHY

FLY ON PAPER FIRST!

FIVE YEAR GOALS

- Elevate Projectile Design and Structural Evaluation Methodology By Including Transient Effects.
- Develop Inexpensive Approach for Construction of Composite Material Components Using Premium Materials.
- Unlock the Secret To Achieving High Performance of Continuous Fiber Composite Under Compression.
- Ply-By-Ply Assessment Technology for Advanced Composite Material Architectures.
- Elevate Rotating Band/Obturator Design From Black Art to an Engineering Science Design Approach.
- Continue Innovative Thought.

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COMPOSITE MATERIALS THRUSTS

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SCOPE:

Particle-Filled Metals - Fracture Toughness Issues.

Continuous Fibers in Metals - Watching Technology.

Particle-Filled Polymers - Some Interest.

Continuous Fibers in Polymers - Prime Interest.

MUST HAVE SOLID COMMERCIAL BASE!

COMPOSITE MATERIALS THRUSTS

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BALLISTIC RESEARCH LABORATORY

THEORETICAL & COMPUTATIONAL -

- Thick Laminated Plate Theory (3D)

- Ply-By-Ply Structural Analysis
  - * Preprocessor
  - * Postprocessor
  - * NIKE/DYNA & ABACUS Compatibility

- Cure Simulation
  - * Integrate Chemistry & Mechanics
  - * Residual Stress Prediction

THEORETICAL ANALYSIS
R. CHRISTENSEN AND S. DETERESA

- NEW CONSTITUTIVE EQUATION
  Developed for fiber composite lamina

$$\boxed{\sigma_{ij} = \lambda \varepsilon_{kk} \delta_{ij} + 2\mu \varepsilon_{ij} + (E_{11} - E)\delta_{1i}\delta_{1j}\varepsilon_{11}}$$

Have used to generate laminate behavior

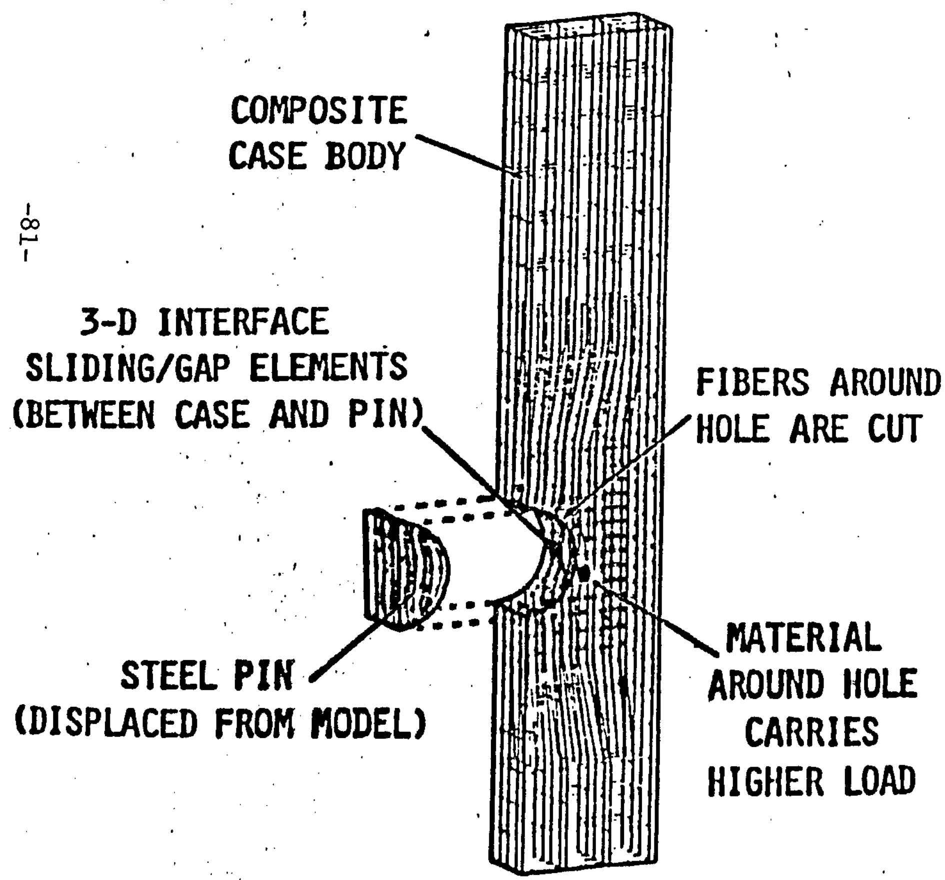
$$\boxed{\sigma_{ij} = \lambda \varepsilon_{kk} \delta_{ij} + 2\mu \varepsilon_{ij} + (E_{11} - E)\sum_{n=1}^{N} m_i^{(n)} m_j^{(n)} m_k^{(n)} m_l^{(n)} \varepsilon_{kl}}$$

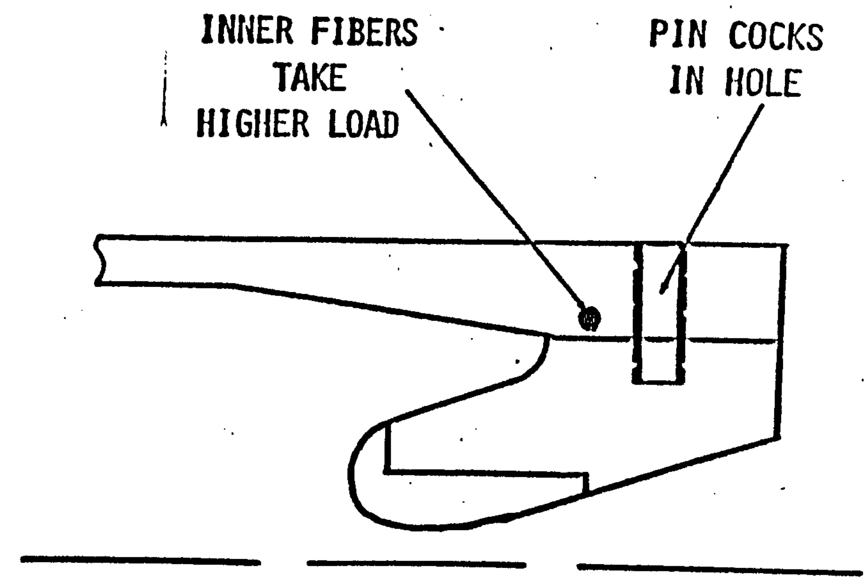
- ADVANTAGES
  Compact tensor formalism
  Fully 3-D applicability
  Possible generalization to nonlinearity

- DISADVANTAGES
  Not user friendly

CUSTOM ANALYTICAL
ENGINEERING SYSTEMS

COMPOSITE PINJOINT TECHNOLOGY

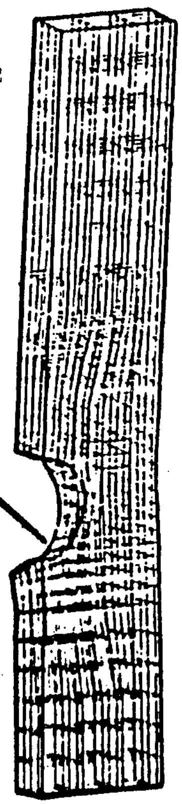




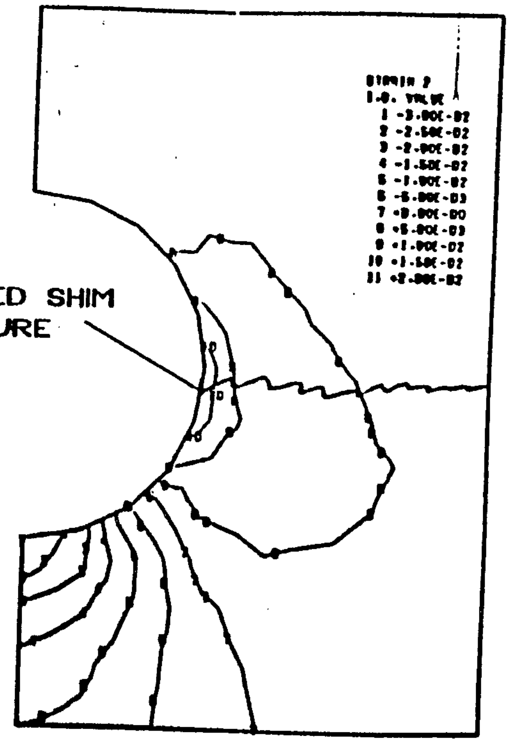
CUSTOM ANALYTICAL
ENGINEERING SYSTEMS

STEEL SHIM REINFORCED
GRAPHITE PINJOINT

PREDICTED FAILURE PRESSURE 6017 PSI
TESTED FAILURE PRESSURE 6022 PSI

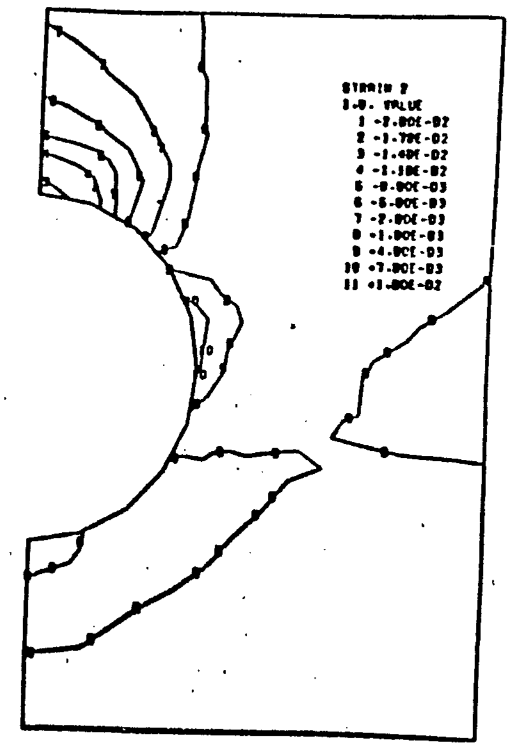


HOLE DISTORTION
FROM
PIN BEARING



PREDICTED SHIM
FAILURE

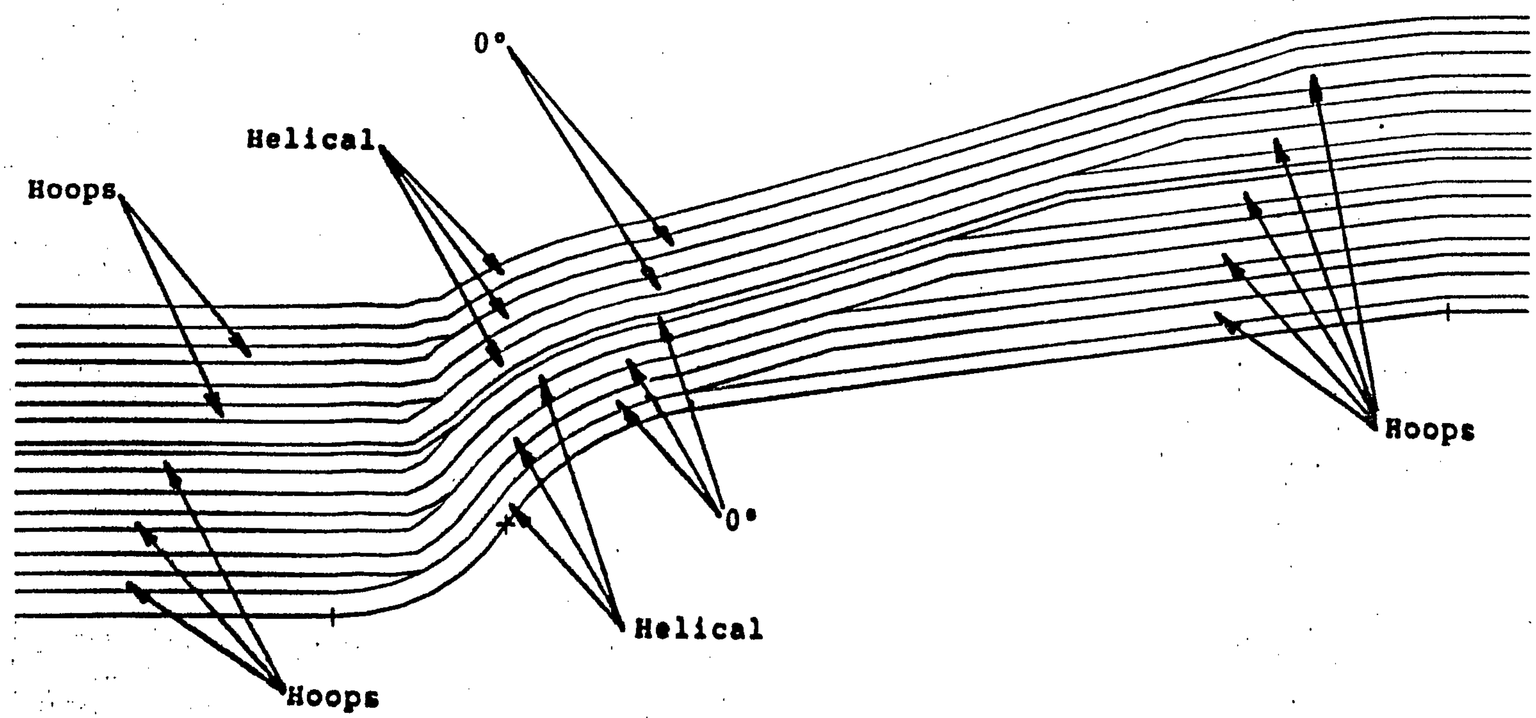
INNER FACE
STRAIN DISTRIBUTION



OUTER FACE
STRAIN DISTRIBUTION

Custom Analytical
Engineering Systems

WOUND-IN CLOSURE LAYUP PATTERN



PROCESSING-PERFORMANCE ISSUES IN THICK THERMOSET COMPOSITES

PROCESSING

- Exotherm Control
- Non-Uniform Cure Gradients
- Winding Tension
- Resin Flow
- Void Content
- Fiber Volume Fraction

PERFORMANCE

- Residual Stresses
- Warpage
- Matrix Cracks
- Delaminations

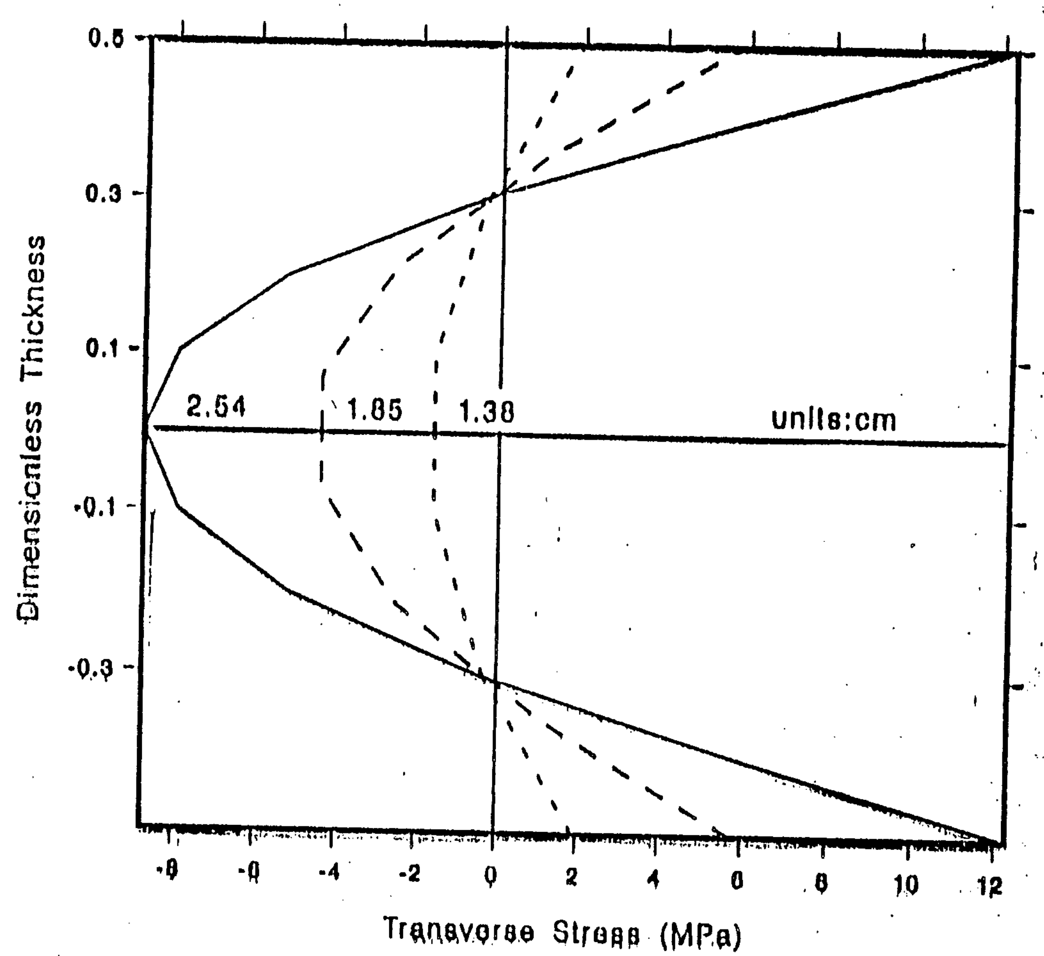
PROCESSING → PERFORMANCE

PROCESS-INDUCED RESIDUAL STRESS MODELING

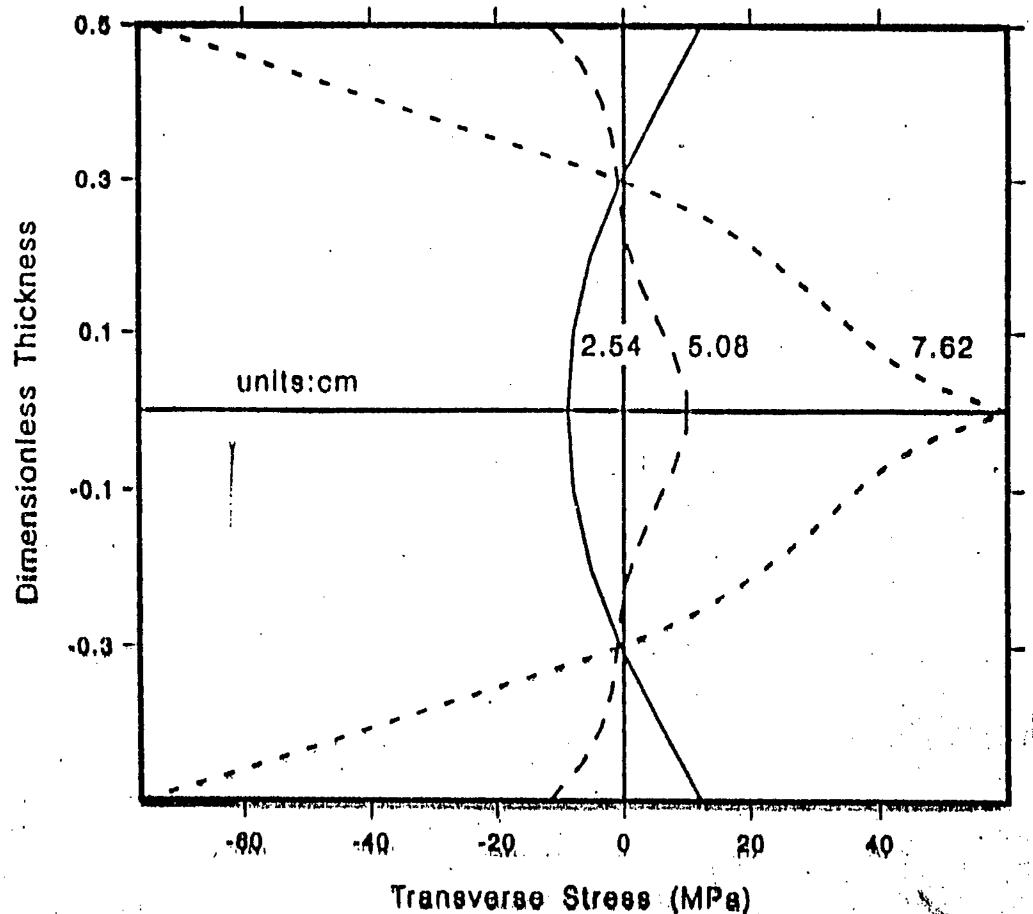
- Heat Transfer
  - heat conduction equation
  - generalized boundary conditions
- Chemical- Kinetics
  - semi-emperical
  - mechanistic
- Cure Shrinkage and Thermal Expansion
  - material models for matrix and fiber
  - self-consistent micromechanics
- Stress Analysis
  - classical laminated plate theory
  - elasticity (generalized plane strain)
  - finite element models

<u>**PREDICT TRANSIENT STRESS DEVELOPMENT DURING CURE**</u>

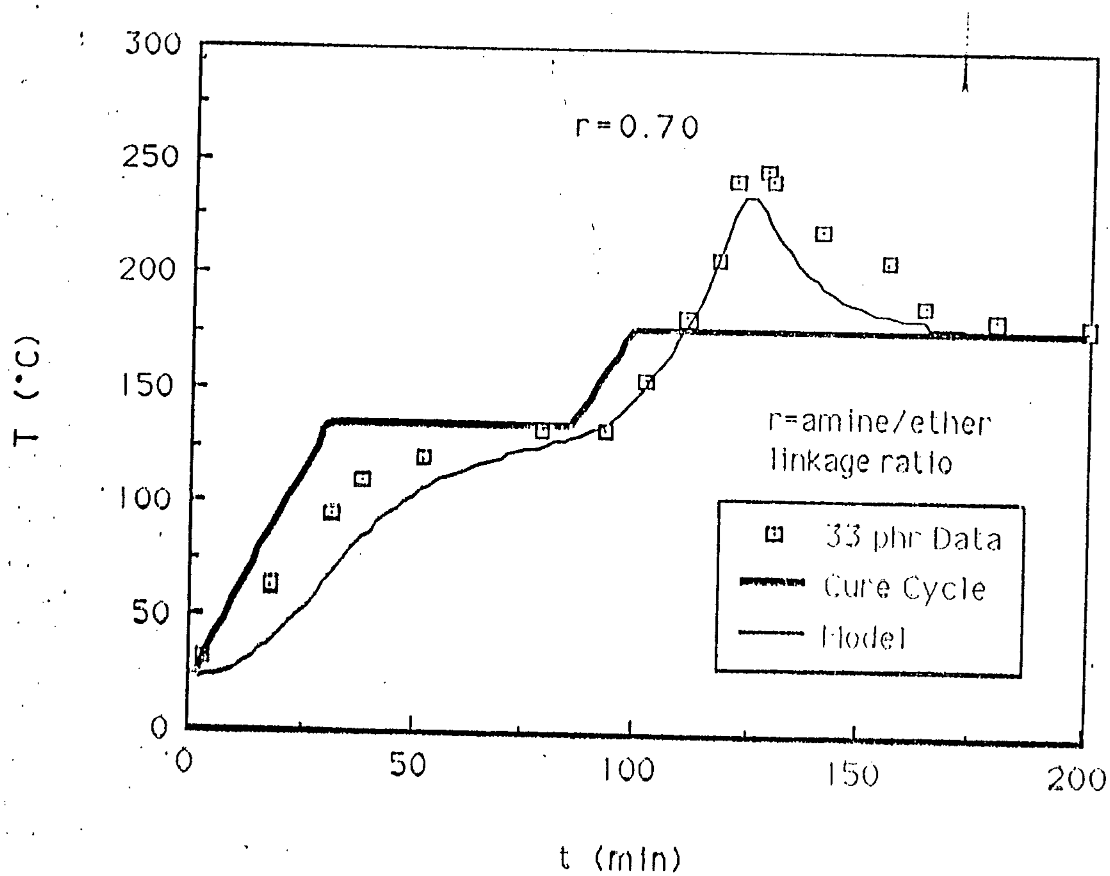
INFLUENCE OF LAMINATE THICKNESS ON RESIDUAL STRESS DEVELOPMENT



INFLUENCE OF LAMINATE THICKNESS ON RESIDUAL STRESS DEVELOPMENT



Mechanistic kinetics used in a laminate cure model, no adjustable parameters



Exp. data of Mijovic *et al.*, 1988

27

FUNDAMENTAL UNDERSTANDING WILL PERMIT INTELLEGENT DESIGN-ENHANCED PERFORMANCE

- How to favorably control residual stress to advantage
- Optimize stacking sequence for application
- Taylor lay-up for end conditions/load transfer through joints
- Section of material systems/combinations (fiber and matrix)

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COMPOSITE MATERIALS THRUSTS

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- Failure Prediction
- Small Effort in Crack Propagation
- Computational Micromechanics
  * Imperfections

PROCESSING -

- Near Net-Shape Fabrication
  *Dry-Fiber Laydown & Resin Transfer Molding
  *Prepreg Tow Laydown

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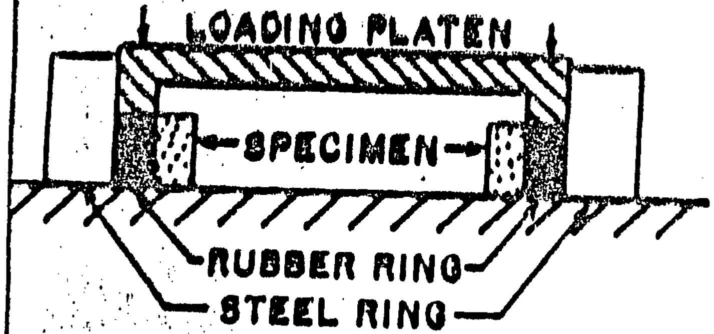
EXPERIMENTAL -

- Development of Multiaxial Tests
- Toward Understanding Compression Strength
  - * Stress/Strain State
  - * Constituents
  - * Architechture
- Material System Studies
- Database
  - *Properties
  - *Interface to FE Codes

ONE-DIMENSIONAL COMPRESSION TEST APPROACH



- Fixture Developed At AFML/URDI (Tsai/Kim)
- Based on Russian Test
- Eliminates Buckling Inherent in Standard (ASTM) Composite Laminate Tests
- Already Adopted by DTRC

RESULTS

- Unidirectional S-Glass/Epoxy-Failure Strength Significantly Greater Than Expected.*
- Unidrectional IM6/Epoxy - Failure Strength Greater Than Expected.*
- Bi-directional S-Glass/Epoxy -Failure Strength Much Greater Than Expected.+
- Bi-directional IM6/Epoxy - Failure Strength as Expected.+

*Compared with available data

+Based on our unidirectional results

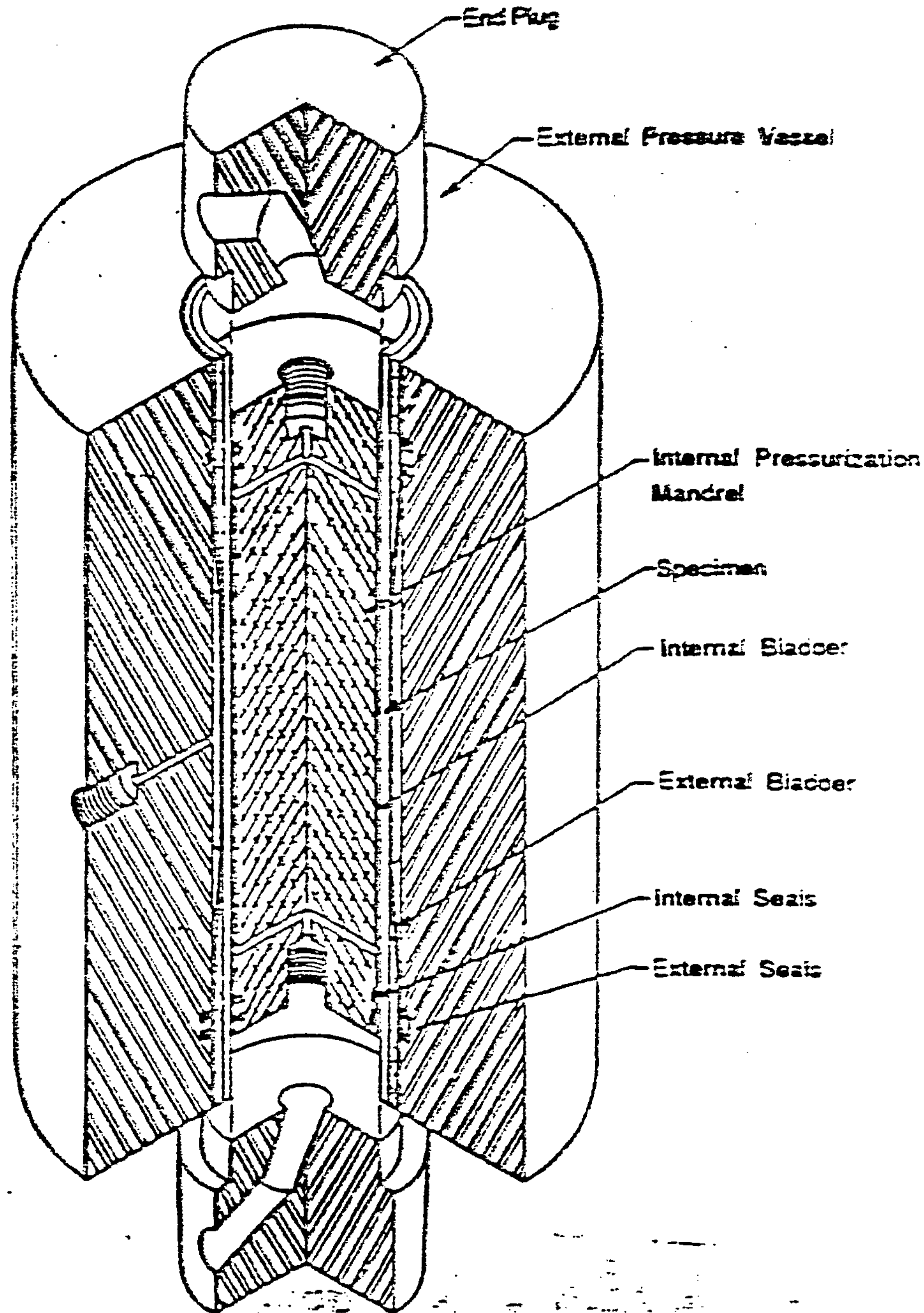


Fig 11. Multiaxial Compression Specimen and Apparatus.

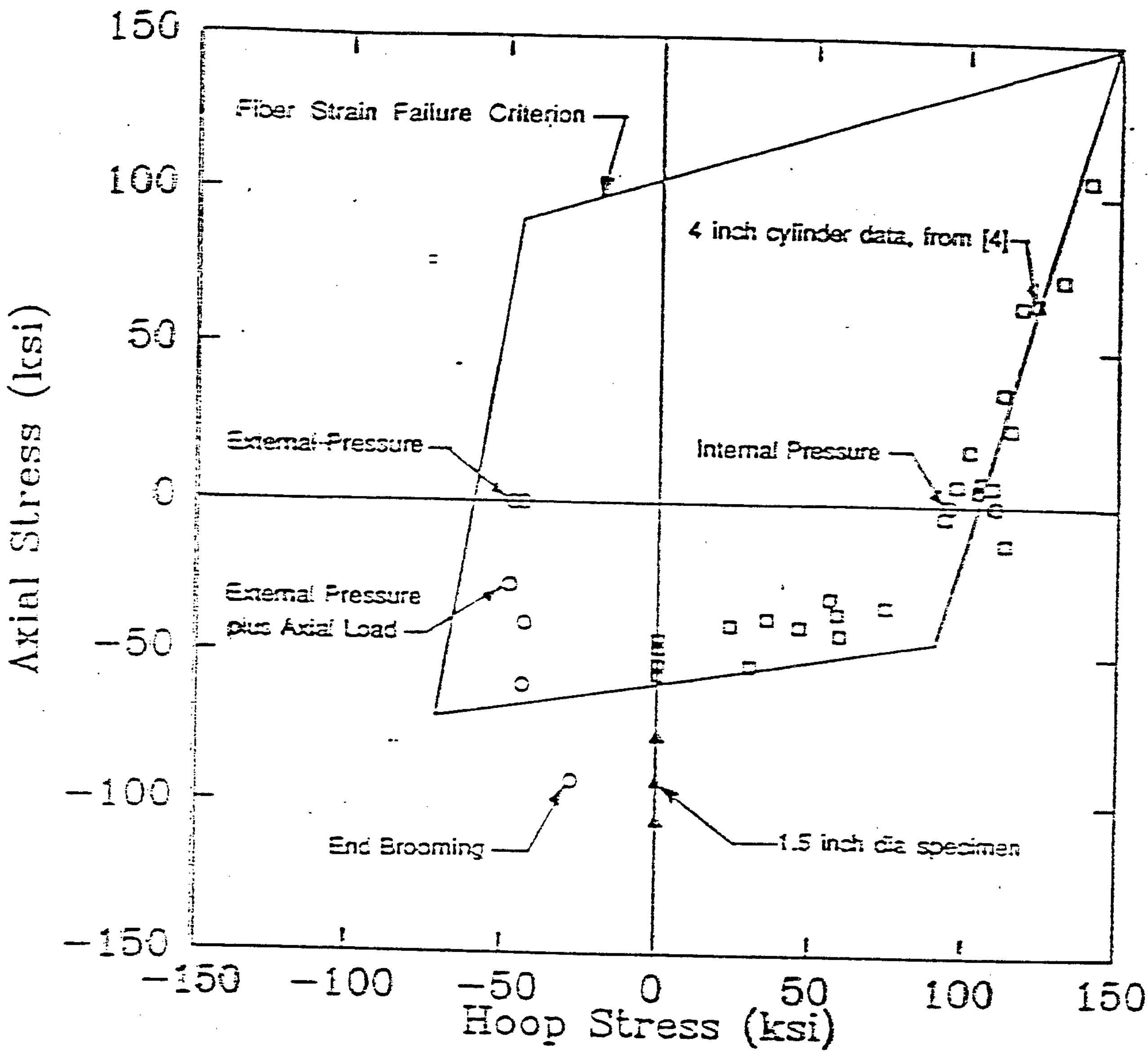
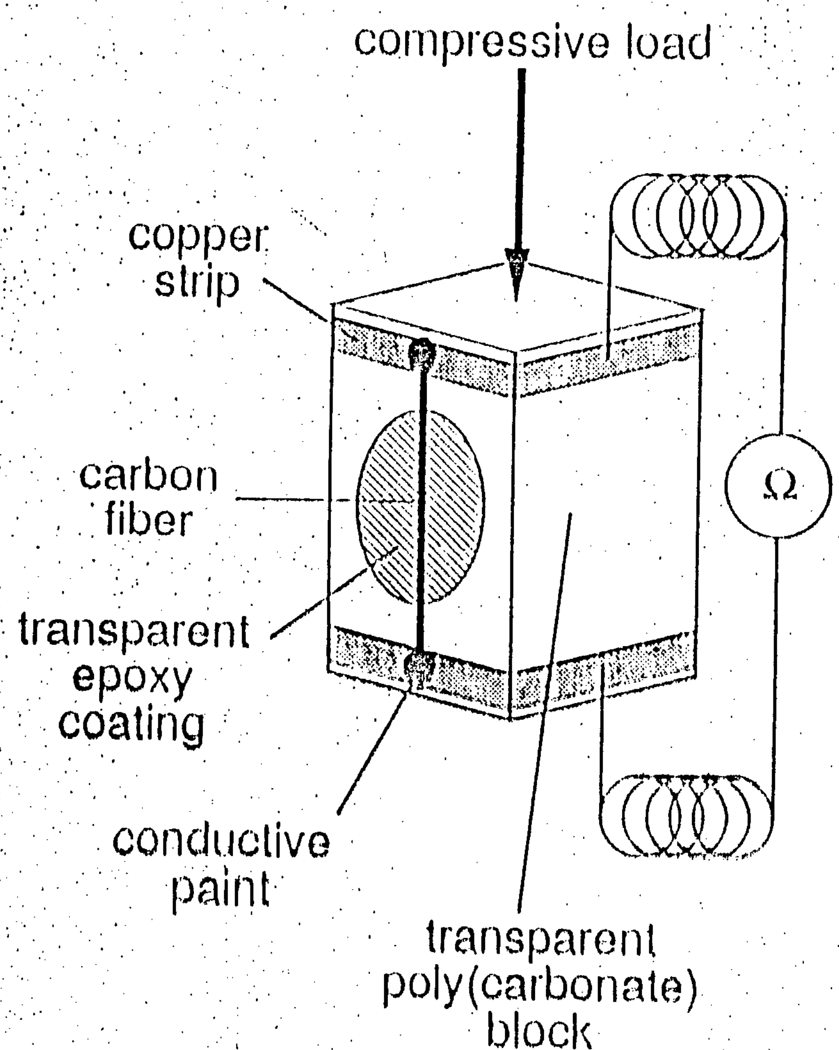
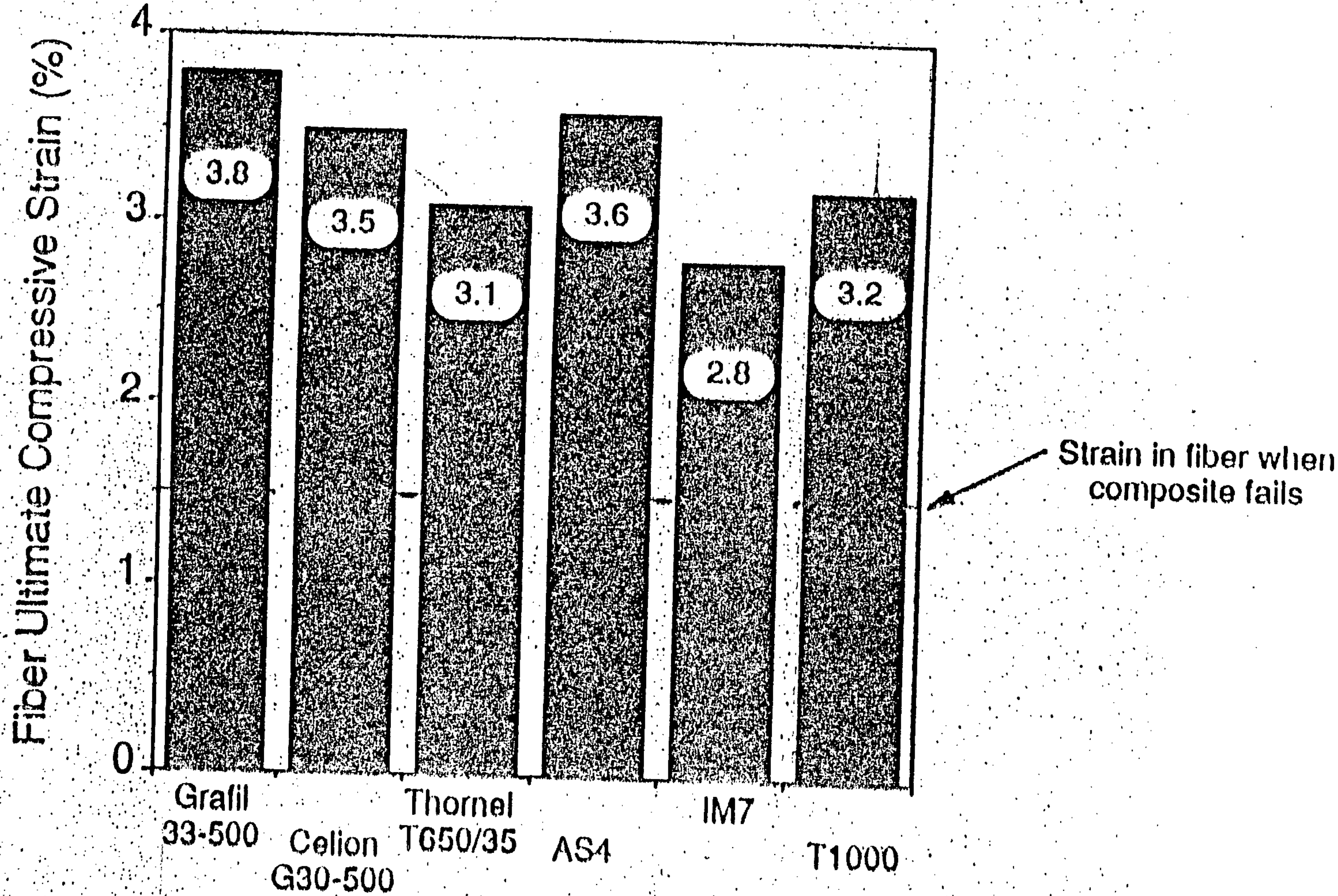


Fig 12. Biaxial Failure Data for $[(90/\pm45/0)_m]_S$ Quasi-Isotropic AS4/3501-6 Carbon/Epoxy Laminates.

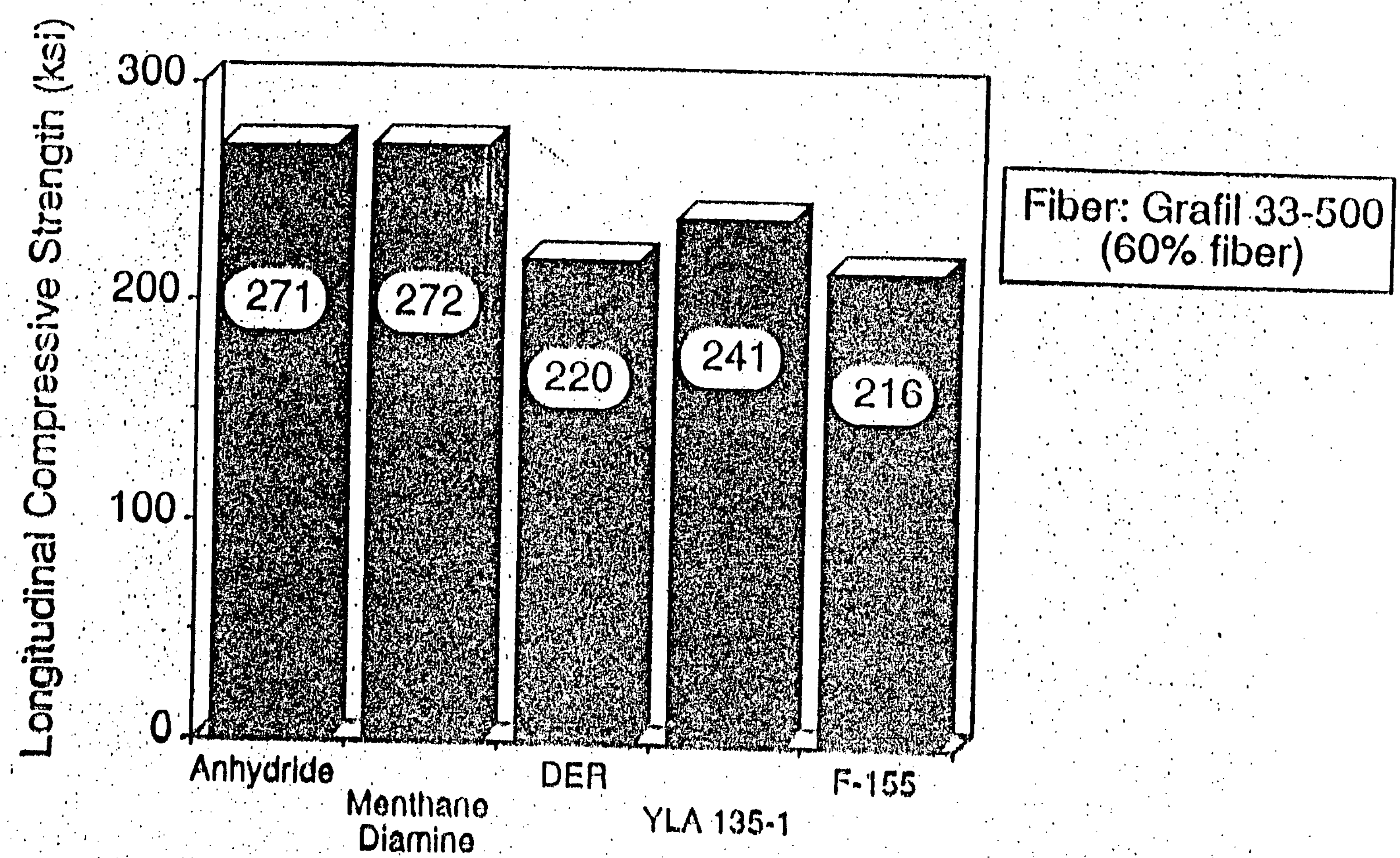
Schematic of Graphite Filament Compression Test



Ultimate Compressive Strains of Graphite Fibers



Longitudinal Compression: Effect of Matrix



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COMPOSITE MATERIAL COMPONENTRY


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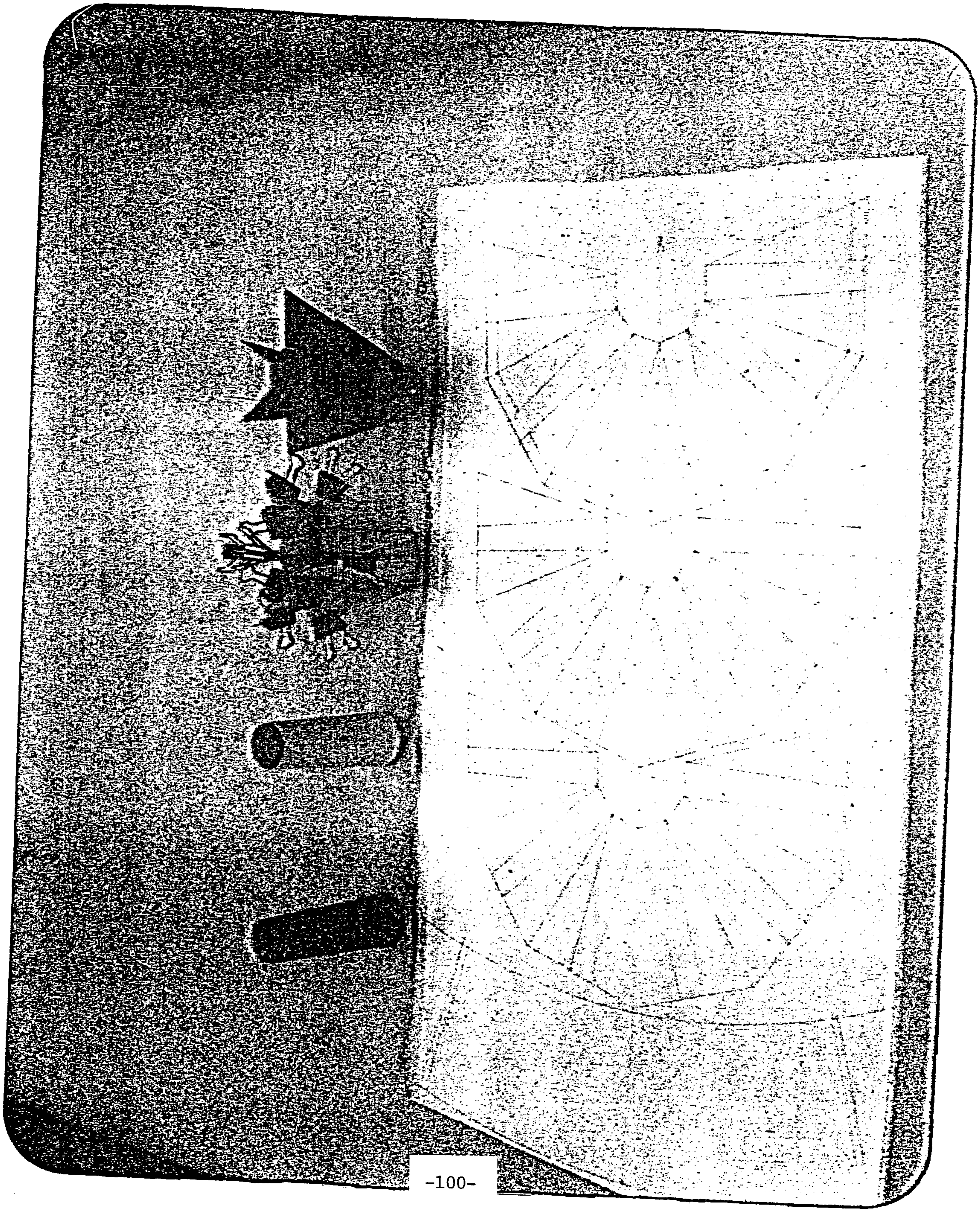

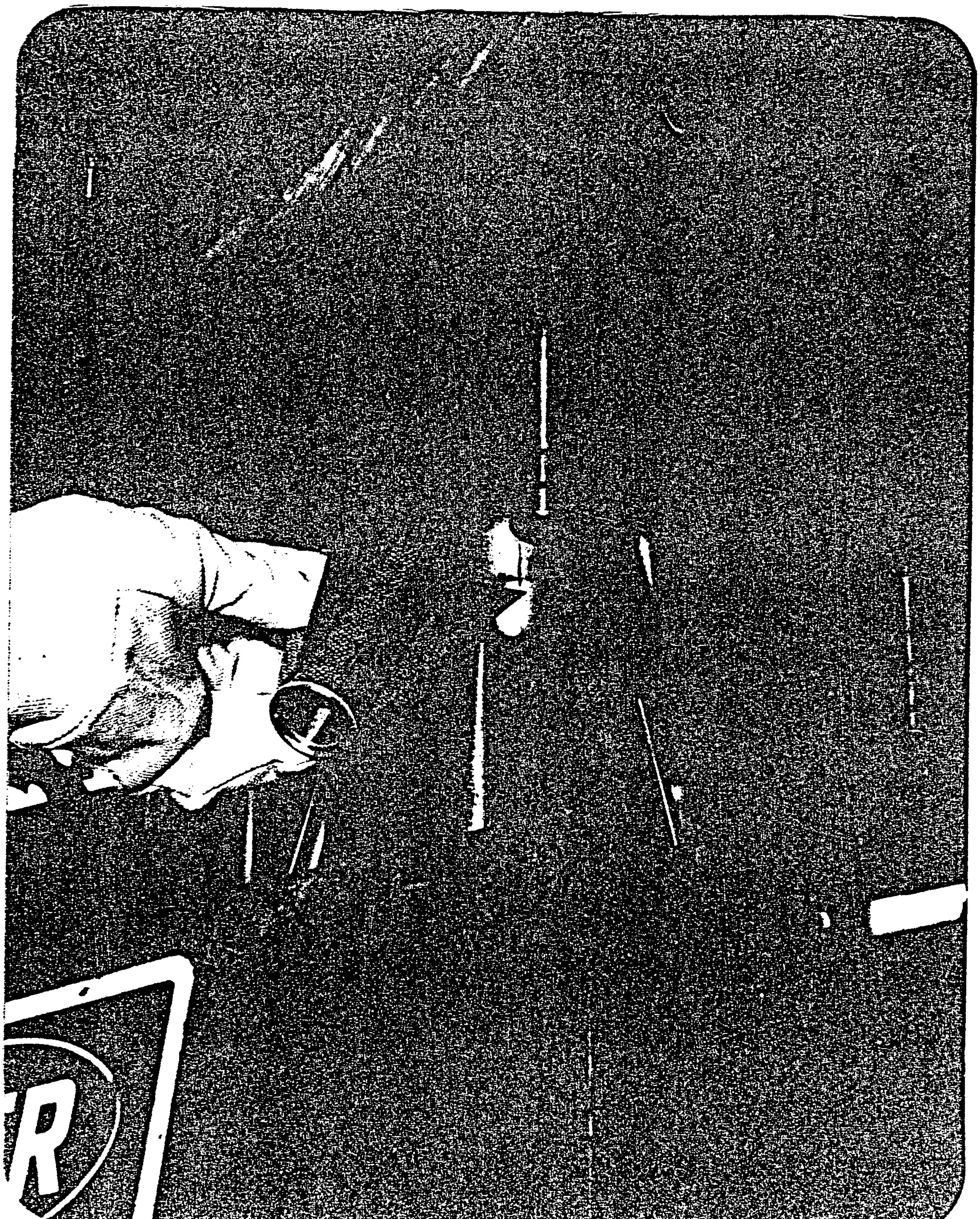

ARMOR TECHNOLOGY

VULNERABILITY METHODOLOGY

US ARMY
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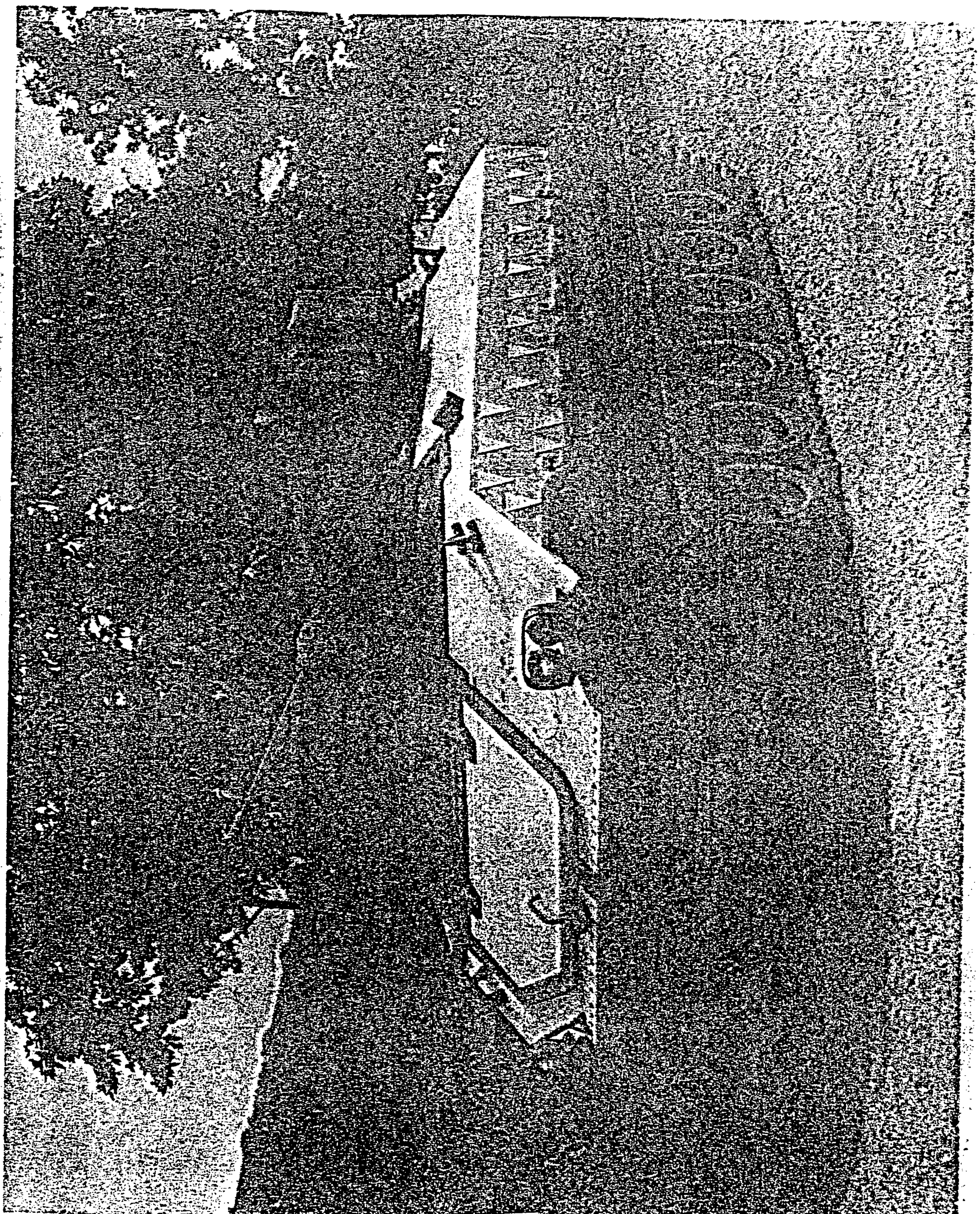
- GROUND VEHICLES
- AIRCRAFT



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VULNERABILITY ASSESSMENT OF COMPOSITE HULL VEHICLES

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OBJECTIVE:

DEVELOP THE ANALYTICAL METHODOLOGY NEEDED TO ASSESS THE VULNERABILITY OF COMPOSITE-HULLED COMBAT VEHICLES TO ATTACK BY CONVENTIONAL MUNITIONS.

MILESTONES:

- ASSEMBLE AND ANALYSE EXISTING DATA RELEVANT TO VULNERABILITY MODELING
- DEVELOP PENETRATION EQUATIONS FOR SHAPED CHARGES AND KE PROJECTILES
- INVESTIGATE THE DAMAGING POTENTIAL OF DEBRIS PRODUCED BY BALLISTIC PERFORATION
- INVESTIGATE THE STRUCTURAL INTEGRITY OF COMPOSITE HULLS AFTER PERFORATION
- INVESTIGATE THE SUSCEPTIBILITY OF COMPOSITE HULLS TO BLAST DAMAGE
- DERIVE ANALYTICAL VULNERABILITY MODELS, ESTIMATE THE VULNERABILITY OF COMPOSITE HULL VEHICLES IN TERMS OF PROBABILITY OF KILL FOR A SPECTRUM OF ATTACKING MUNITIONS

CATASTROPIC STRUCTURAL KILL ANALYSIS

- Effect of Damage to a Composite A/C Horizontal Stabilizer on Aeroelastic Stability.

- Test Code (MSC/NASTRAN) Usefulness - Validation of Methodology is the Issue.

BRL WRAP-UP

- Multidirectional Compression Offers Many Challenging Aspects.
- Computational Models of Thick Structures - Must Back Out Accurate Assessments of Ply Stresses/Strains.
- Subcritical National Effort for Constitutive/Failure Behavior at High Loading Rates (10 to 5000/sec).
- Future Funding Uncertainties - Critical for Particular Programs.

BRL CAPABILITIES/EXPERTISE

CAPABILITIES

Advanced Simulation
Supercomputing
Mechanical Testing Machines
Scanning Electron Microscopy
High Rate Tester
Limited Sample Fabrication Capability
Chemical Analysis

EXPERTISE

Ballistic Technologies
Design
Dynamic & Transient Structural Analysis
Development of Prototypes
Ballistic Testing
Vulnerablity Assessment
Residual Stress Prediction
Growing Expertise in:
- Processing
- Material Characterization

Curing Stresses In Thick Thermoset Laminates

Travis A. Bogetti
Mechanics & Structures Branch
U.S. Army Ballistic Research Lab
Aberdeen Proving Ground, MD

RESEARCH GOAL EVOLVED FROM LITERATURE REVIEW IN THICK-SECTION COMPOSITES:

MECHANICS OF THICK SECTIONS?

Problem Areas:

- Processing
- Testing
- Analysis

Poor Laminate Quality:

- Matrix cracks
- Delaminations
- Warpage

PROCESSING → PERFORMANCE

SOURCES OF RESIDUAL STRESS

Microscopic

Fiber/Matrix
- Thermal Expansion
- Chemical Shrinkage

Heterogeneity

Macroscopic

Stacking Sequence
- Thermal Expansion
- Chemical Shrinkage

Processing

Thermal Gradients

Time-Dependent Properties
- Degree of Cure
- Viscoelastic

Material Point — Self-Equilibrating — Structure

RESEARCH GOAL

Relate the processing history of thick-section thermosetting composites to their performance characteristics in terms of overall part quality, dimensional stability and state of residual stress.

APPROACH

- Develop a methodology to study process induced stresses and deformations in thick-section composites accounting for:
  - Thermo-chemical interactions.
  - Thermal and chemical volume changes.
  - Kinetic-viscoelastic material behavior.
  - Spatial variations.
- Verify model predictions with experimental observations.

OVERVIEW

- Introduction
- $\boxed{\text{Cure Simulation Analysis/Results}}$
- Material Models
- Stress Analysis/Results
- Conclusions
- Current Activities/Future Work

CURE SIMULATION NUMERICAL ANALYSIS

- Governing equations for thermo-chemical interactions.
- Boundary-fitted coordinate system (BFCS) grid generation and transformation technique.
- Alternating direction explicit (ADE) finite difference solution technique.

GOVERNING EQUATIONS

Heat transfer:

$$\dot{q} + k_{xx}\frac{\partial^2 T}{\partial x^2} + 2k_{xz}\frac{\partial^2 T}{\partial x \partial z} + k_{zz}\frac{\partial^2 T}{\partial z^2} = \rho c \frac{\partial T}{\partial t}$$

$$a\frac{\partial T}{\partial \hat{n}} + bT + c = 0 \quad \text{(boundary condition)}$$

Chemical Kinetics:

$$\dot{q} = \rho H_r \left(\frac{d\alpha}{dt} \right)$$

$$\alpha = \int_0^t \left(\frac{d\alpha}{dt} \right) dt$$

DEGREE OF CURE MODEL FOR GLASS/POLYESTER

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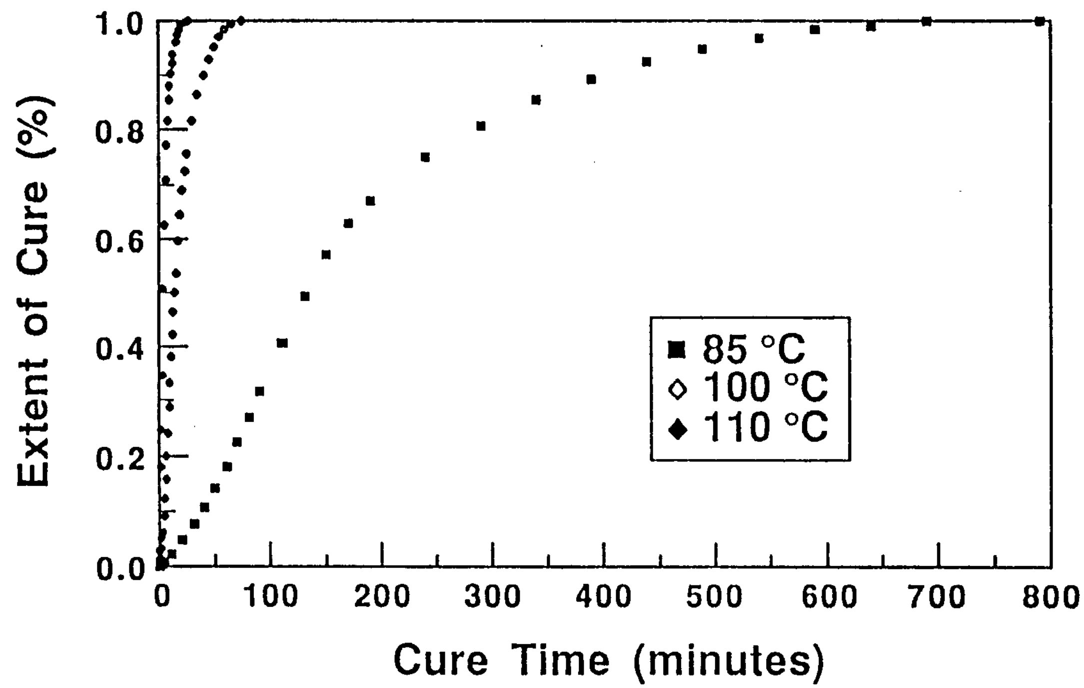

Reference: Adams

EXPERIMENTAL VERIFICATION OF CURE SIMULATION ANALYSIS

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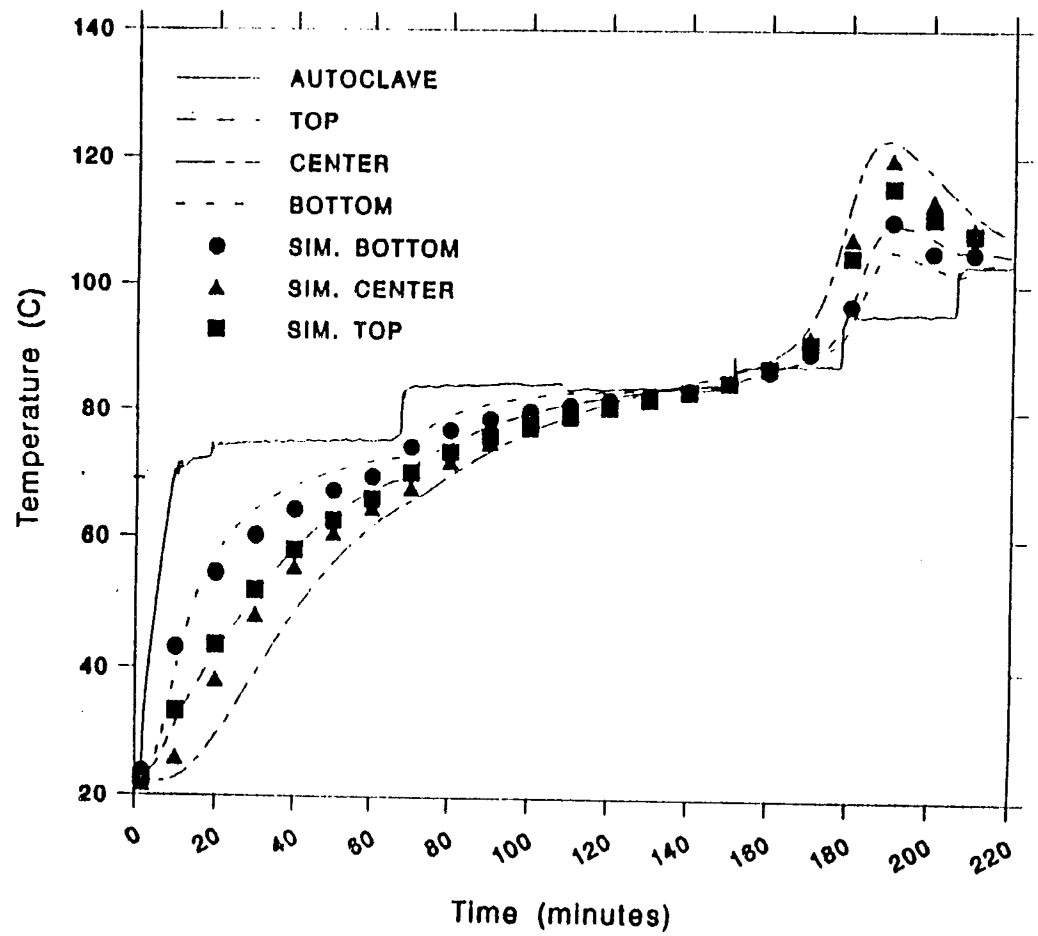

TEMPERATURE CONTOUR IN GLASS/POLYESTER ANGLE BEND AT EXOTHERM

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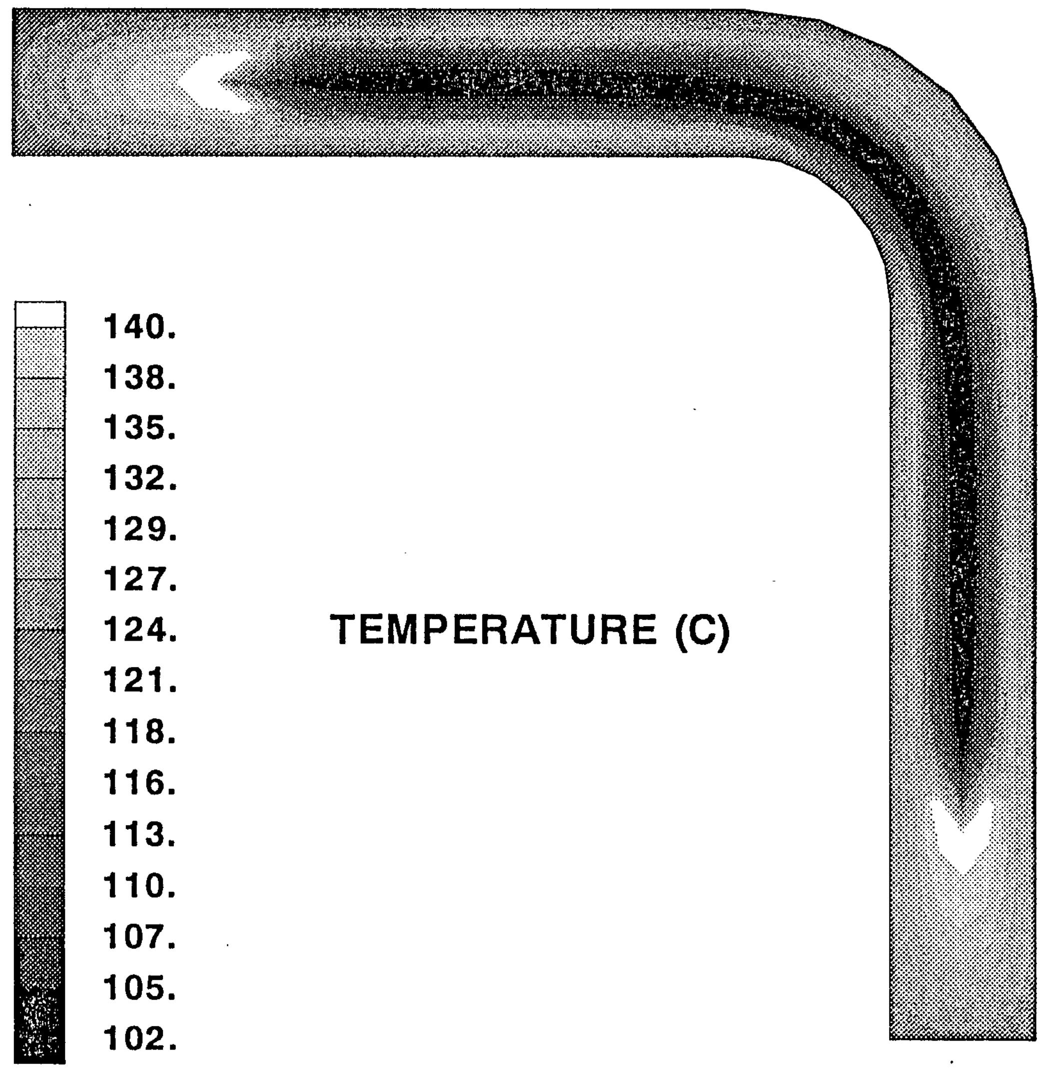

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DEGREE OF CURE CONTOUR IN GLASS/POLYESTER ANGLE BEND AT EXOTHERM

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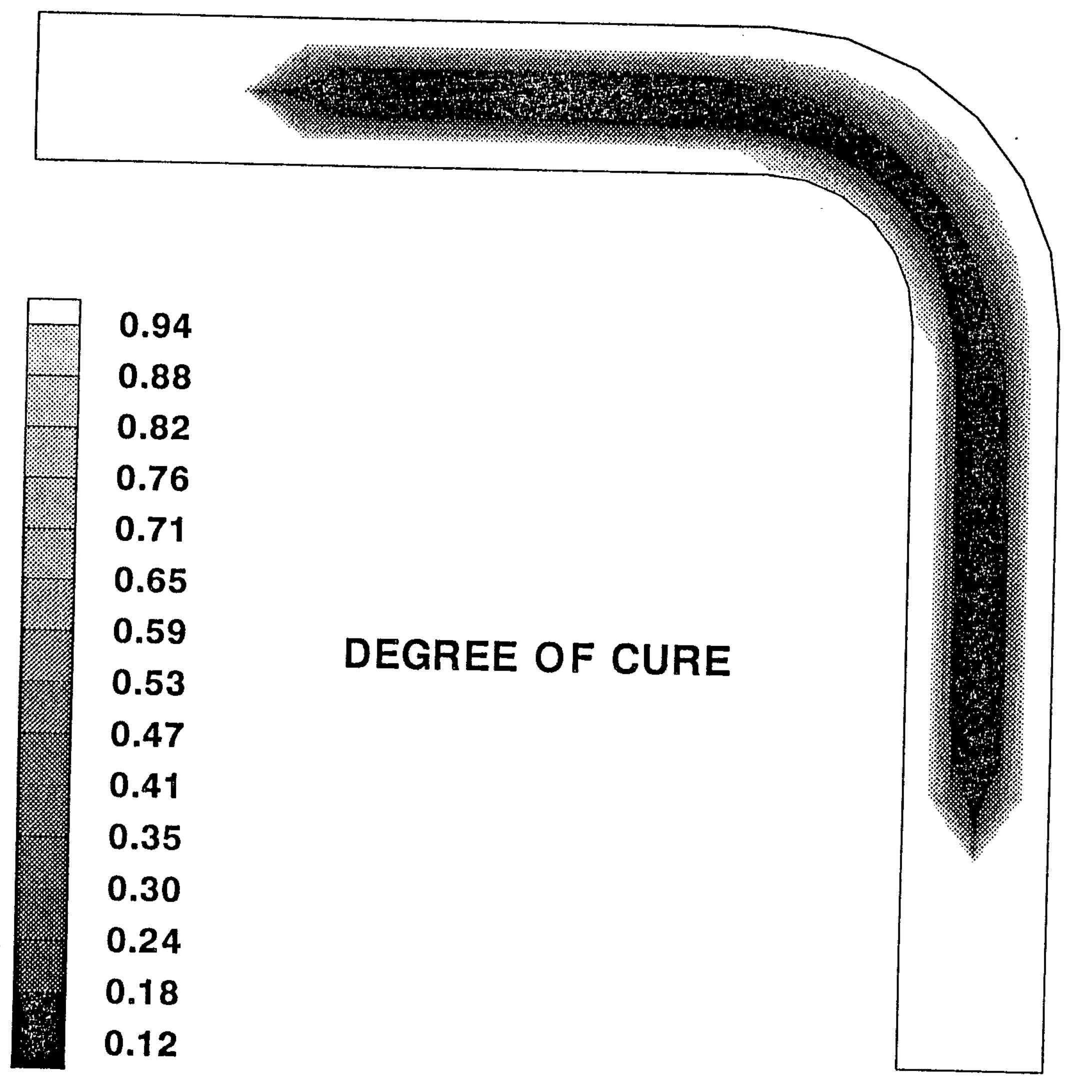

OVERVIEW

- Introduction
- Cure Simulation Analysis/Results
- Material Models
- Stress Analysis/Results
- Conclusions
- Current Activities/Future Work

MATERIAL MODELS

- Resin Modulus
- Chemical Shrinkage Strain
- Thermal Expansion Strain
- Micromechanics

THERMOSET RESIN BEHAVIOR DURING CURE

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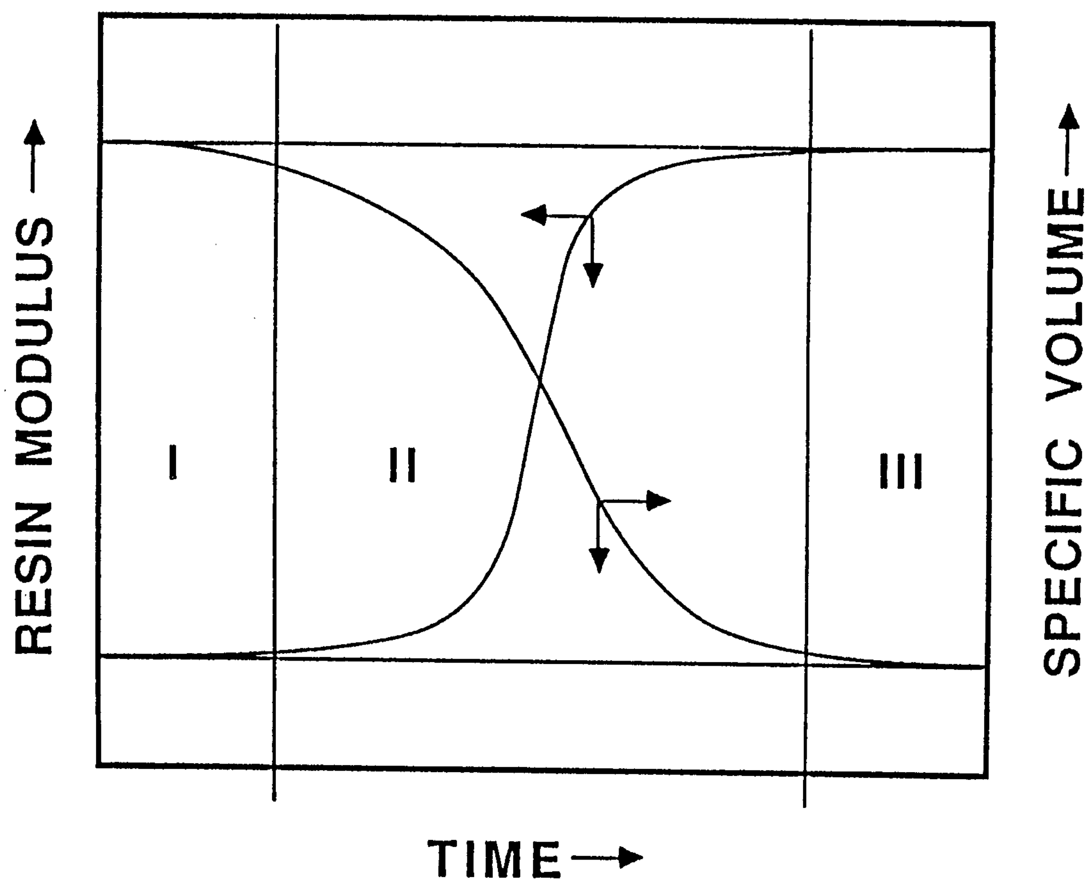

MATERIAL MODEL ASSUMPTIONS

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RESIN:

Moduli = f(α)

Thermal Expansion = constant

Cure Shrinkage = f(α)

FIBER:

Moduli = constant

Thermal Expansion = constant

Cure Shrinkage = 0

MICROMECHANICS → Effective Homogeneous Composite Properties and Strains

POLYESTER RESIN MODULI DURING ISOTHERMAL CURE AT 100 C

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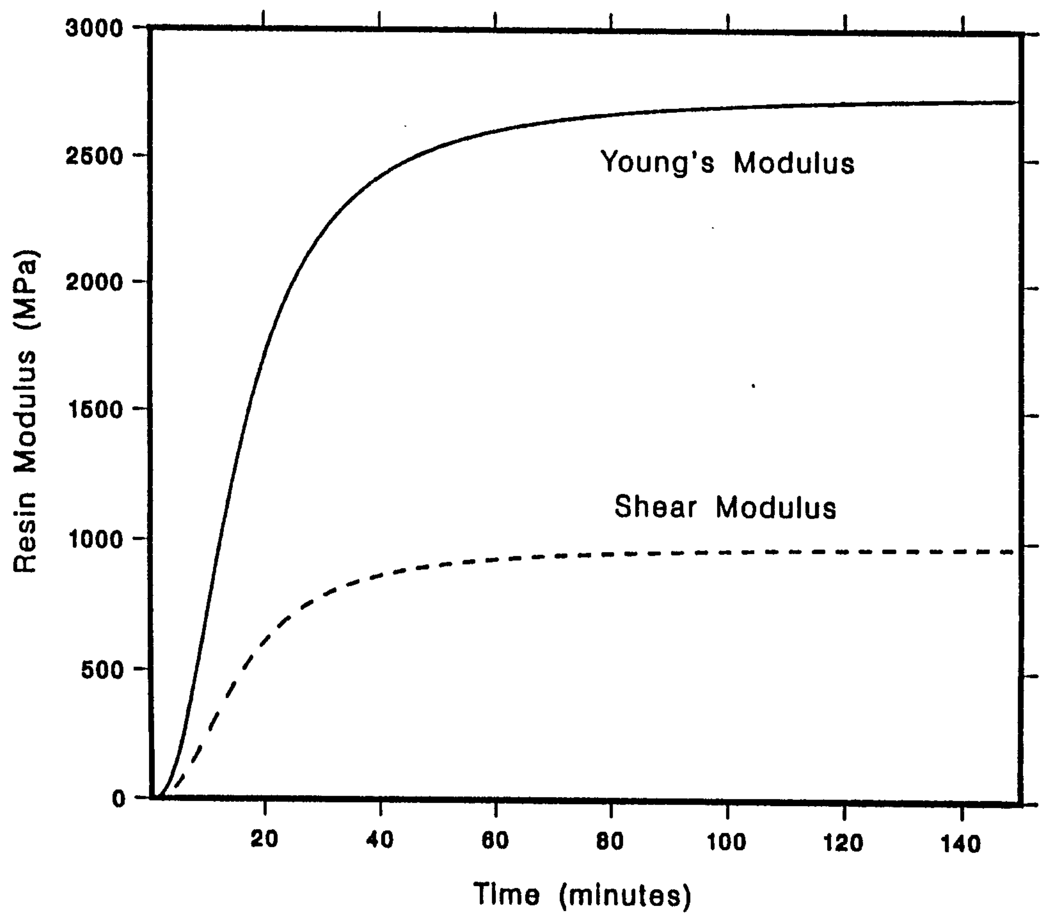

EFFECTIVE COMPOSITE PROPERTIES DURING ISOTHERMAL CURE AT 100 C

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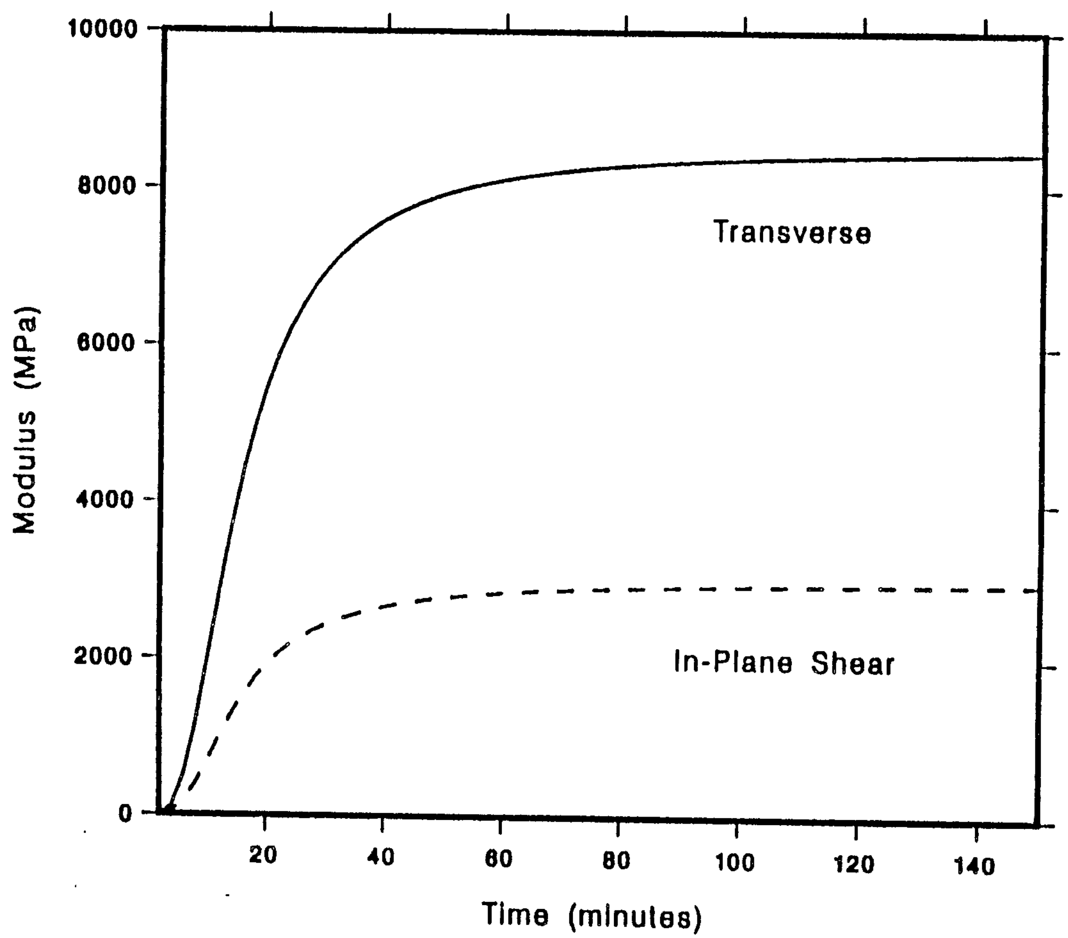

CHEMICAL SHRINKAGE STRAIN DURING A TYPICAL CURE CYCLE

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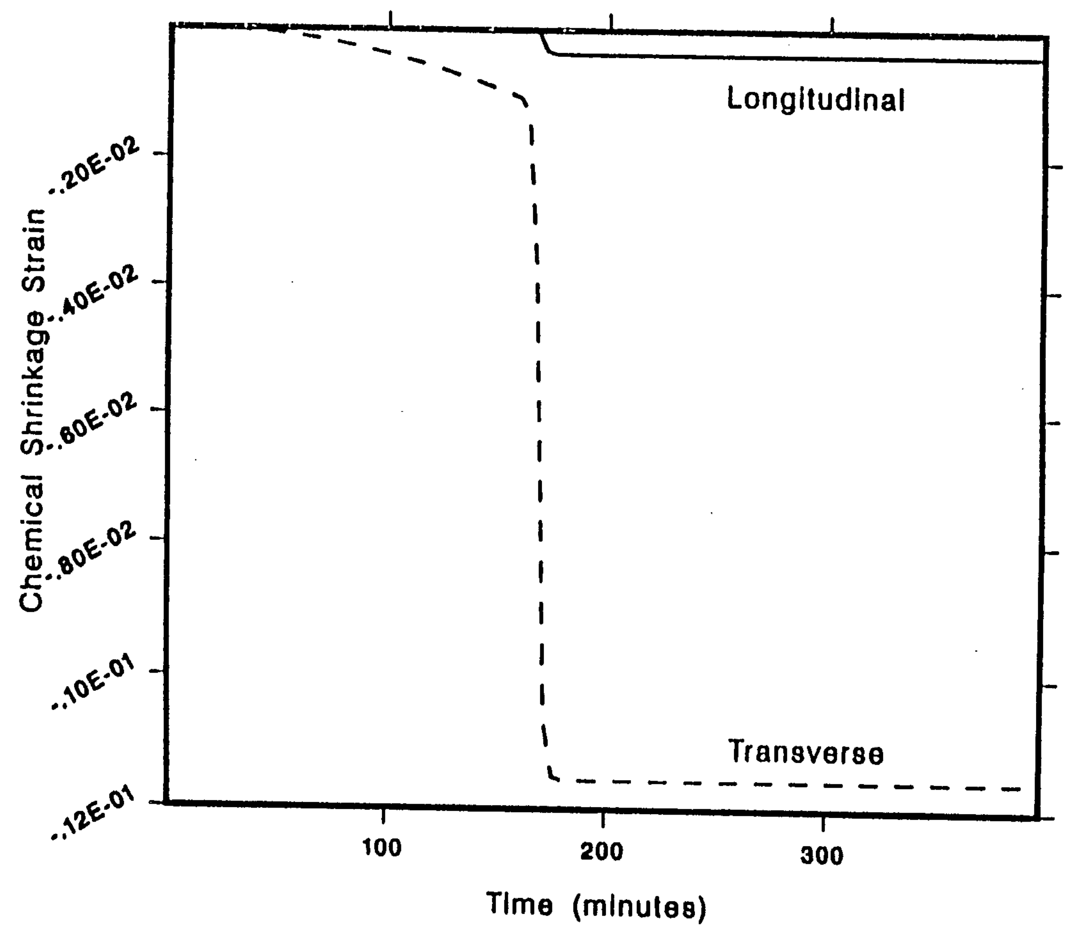

THERMAL EXPANSION STRAIN DURING A TYPICAL CURE CYCLE

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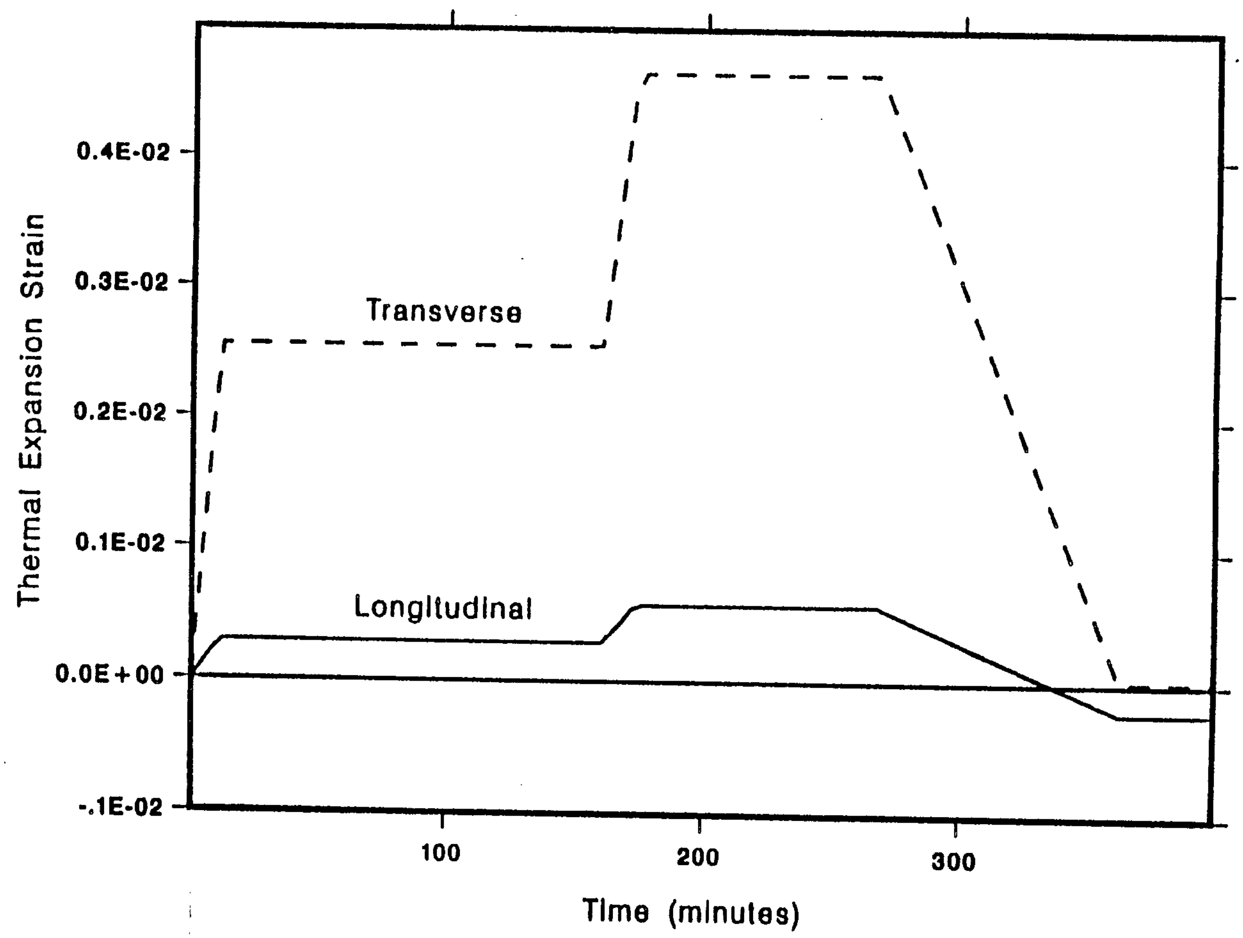

TOTAL PROCESS-INDUCED STRAIN DURING A TYPICAL CURE CYCLE

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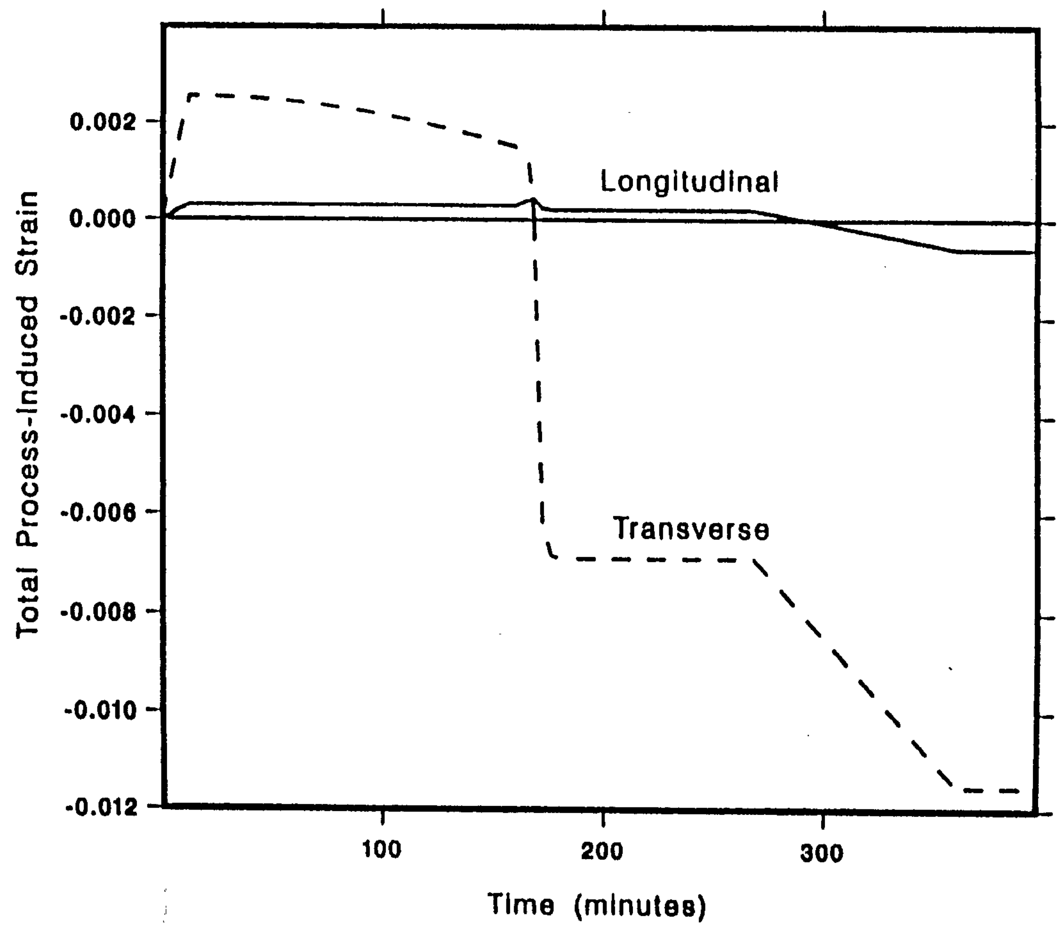

OVERVIEW

- Introduction
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PROCESS-INDUCED INCREMENTAL STRESS ANALYSIS

Effective In-Plane Force and Moment Resultants:

$$(\Delta N_x^T, \Delta M_x^T) = \sum_{k=1}^{N} (\bar{Q}_{11}\Delta\epsilon_x^T + \bar{Q}_{12}\Delta\epsilon_y^T + \bar{Q}_{16}\Delta\epsilon_{xy}^T)(1, z)$$

$$(\Delta N_y^T, \Delta M_y^T) = \sum_{k=1}^{N} (\bar{Q}_{12}\Delta\epsilon_x^T + \bar{Q}_{22}\Delta\epsilon_y^T + \bar{Q}_{26}\Delta\epsilon_{xy}^T)(1, z)$$

$$(\Delta N_{xy}^T, \Delta M_{xy}^T) = \sum_{k=1}^{N} (\bar{Q}_{16}\Delta\epsilon_x^T + \bar{Q}_{26}\Delta\epsilon_y^T + \bar{Q}_{66}\Delta\epsilon_{xy}^T)(1, z)$$

Effective Laminate Deformations:

$$\begin{Bmatrix} \Delta\epsilon_x^o \\ \Delta\epsilon_y^o \\ \Delta\epsilon_{xy}^o \\ \Delta\kappa_x^o \\ \Delta\kappa_y^o \\ \Delta\kappa_{xy}^o \end{Bmatrix} = \begin{bmatrix} a_{11} & a_{12} & a_{16} & b_{11} & b_{12} & b_{16} \\ a_{12} & a_{22} & a_{26} & b_{12} & b_{22} & b_{26} \\ a_{16} & a_{26} & a_{66} & b_{16} & b_{26} & b_{66} \\ b_{11} & b_{12} & b_{16} & d_{11} & d_{12} & d_{16} \\ b_{12} & b_{22} & b_{26} & d_{12} & d_{22} & d_{26} \\ b_{16} & b_{26} & b_{66} & d_{16} & d_{26} & d_{66} \end{bmatrix} \begin{Bmatrix} \Delta N_x^T \\ \Delta N_y^T \\ \Delta N_{xy}^T \\ \Delta M_x^T \\ \Delta M_y^T \\ \Delta M_{xy}^T \end{Bmatrix}$$

PROCESS-INDUCED INCREMENTAL STRESS ANALYSIS (CONT'D)

Strain-Displacement Relations:

$$\Delta\epsilon_x = \Delta\epsilon_x^o + z\Delta\kappa_x$$

$$\Delta\epsilon_x = \Delta\epsilon_y^o + z\Delta\kappa_y$$

$$\Delta\epsilon_{xy} = \Delta\epsilon_{xy}^o + z\Delta\kappa_{xy}$$

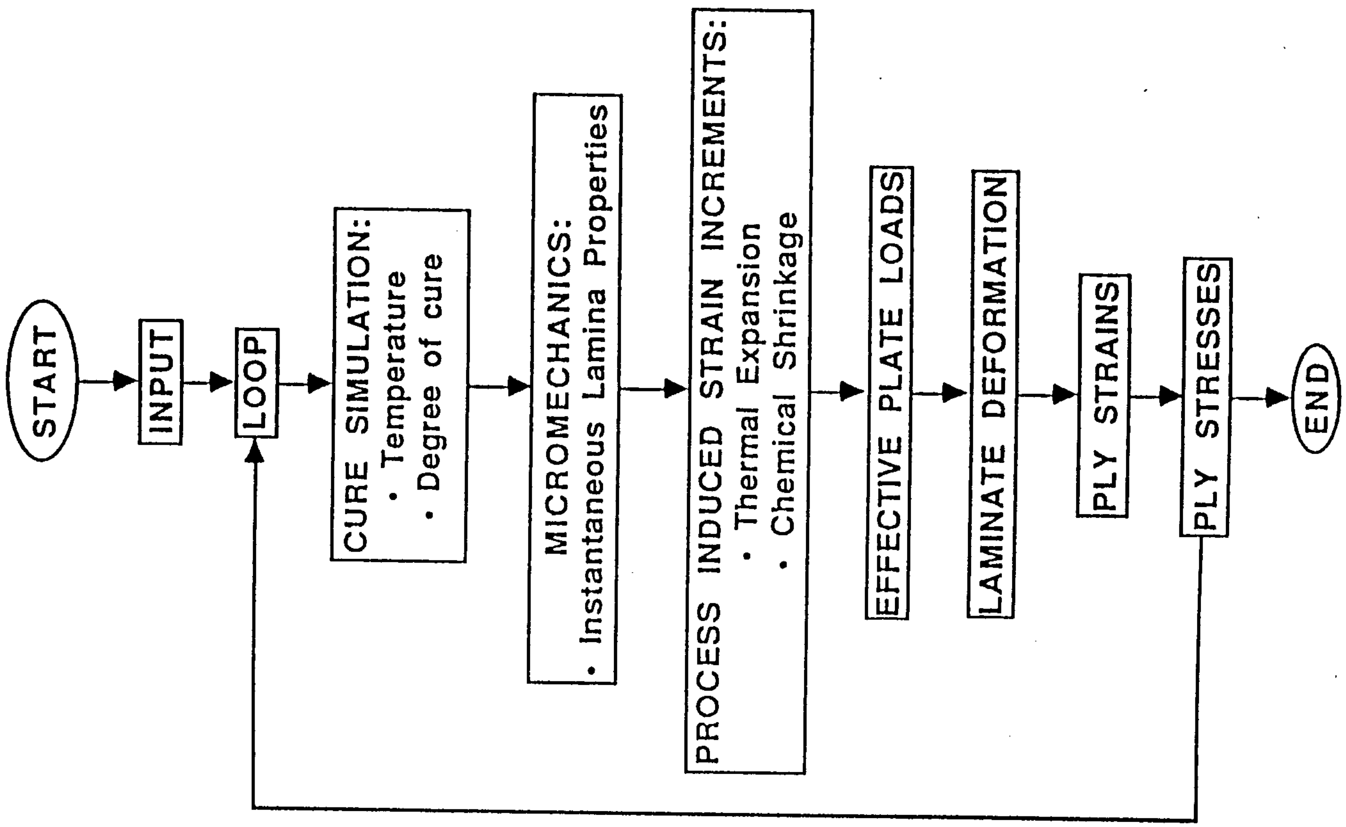
Ply Stresses:

$$\Delta\sigma_x = \bar{Q}_{11}(\Delta\epsilon_x - \Delta\epsilon_x^T) + \bar{Q}_{12}(\Delta\epsilon_y - \Delta\epsilon_y^T) + \bar{Q}_{16}(\Delta\epsilon_{xy} - \Delta\epsilon_{xy}^T)$$

$$\Delta\sigma_y = \bar{Q}_{12}(\Delta\epsilon_x - \Delta\epsilon_x^T) + \bar{Q}_{22}(\Delta\epsilon_y - \Delta\epsilon_y^T) + \bar{Q}_{26}(\Delta\epsilon_{xy} - \Delta\epsilon_{xy}^T)$$

$$\Delta\sigma_{xy} = \bar{Q}_{16}(\Delta\epsilon_x - \Delta\epsilon_x^T) + \bar{Q}_{26}(\Delta\epsilon_y - \Delta\epsilon_y^T) + \bar{Q}_{66}(\Delta\epsilon_{xy} - \Delta\epsilon_{xy}^T)$$

ANALYSIS METHODOLOGY



NORMALIZED TRANSVERSE MODULUS PREDICTIONS COMPARED TO EXPERIMENTAL MEASUREMENTS IN GR/EP (Hahn and Kim)

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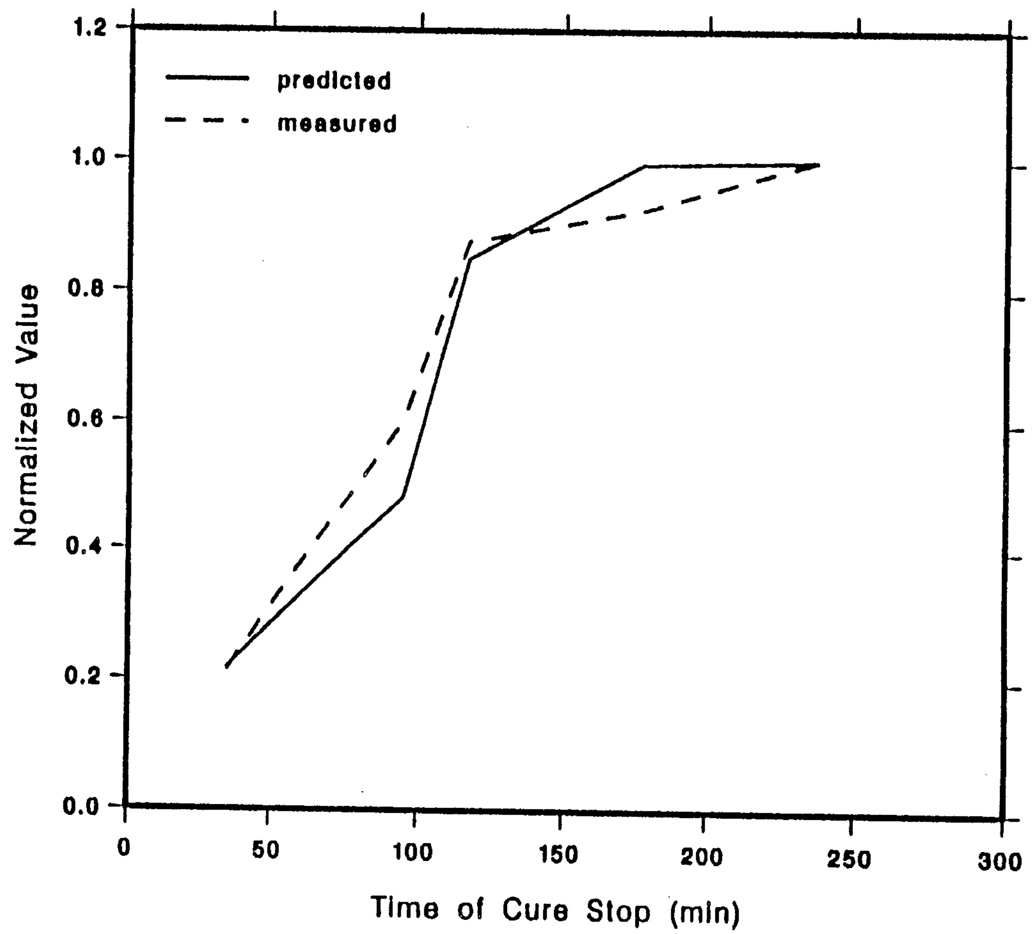

CURVATURE PREDICTIONS COMPARED TO EXPERIMENTAL MEASUREMENTS IN A [0/90] GR/EP LAMINATE (Hahn and Kim)

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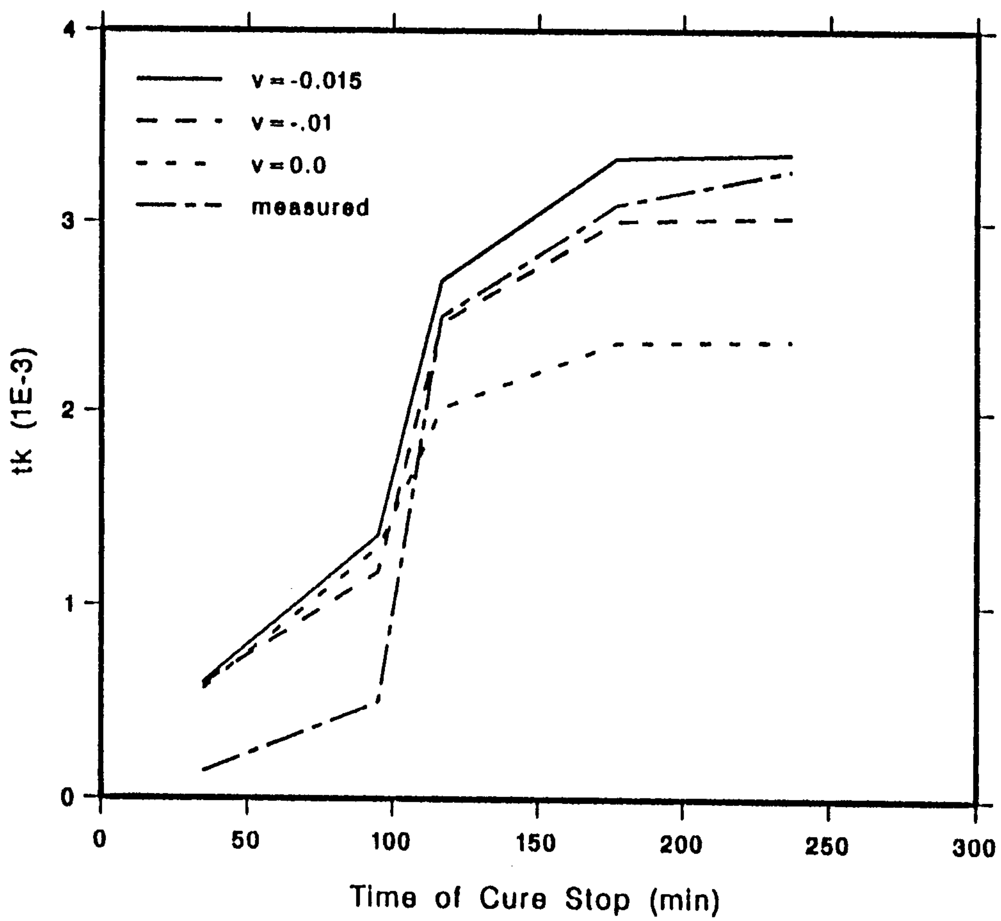

TYPICAL GLASS/POLYESTER TEMPERATURE CURE CYCLE

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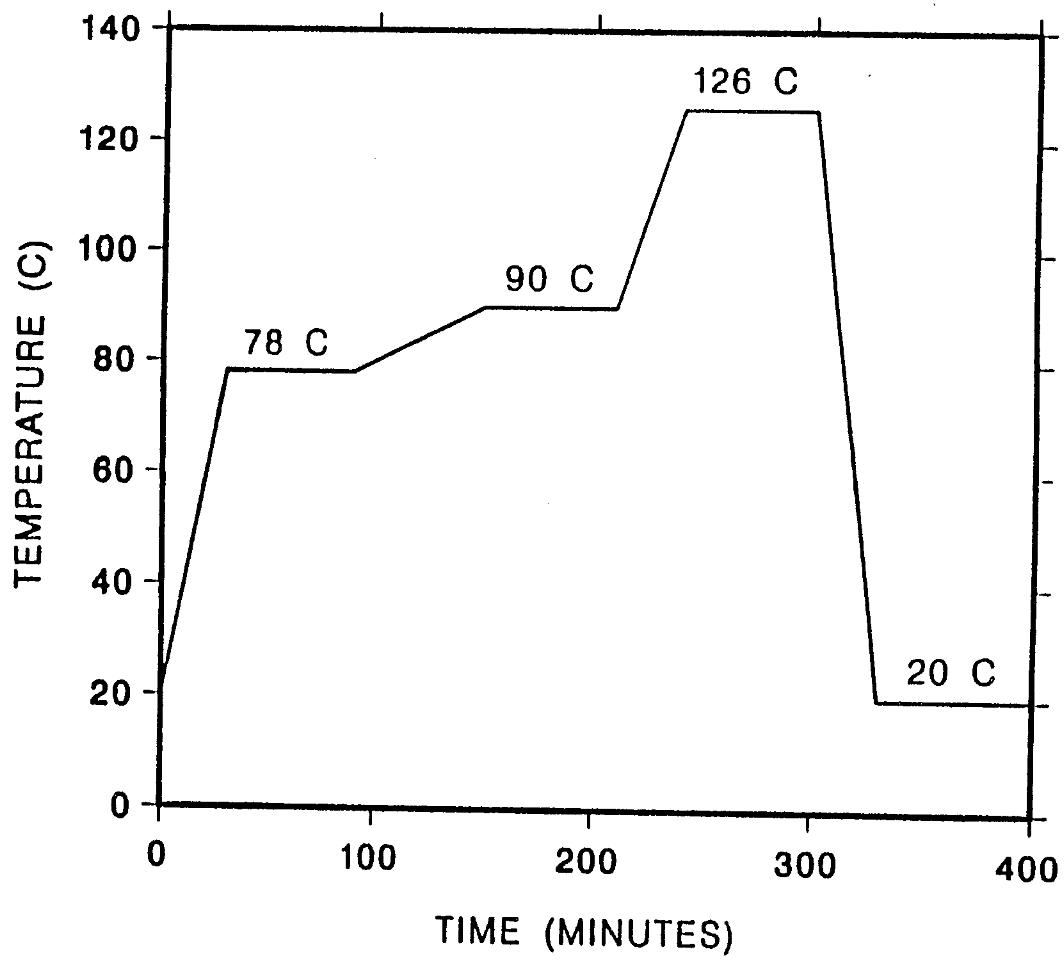

TEMPERATURE GRADIENTS AT 164 MINUTES

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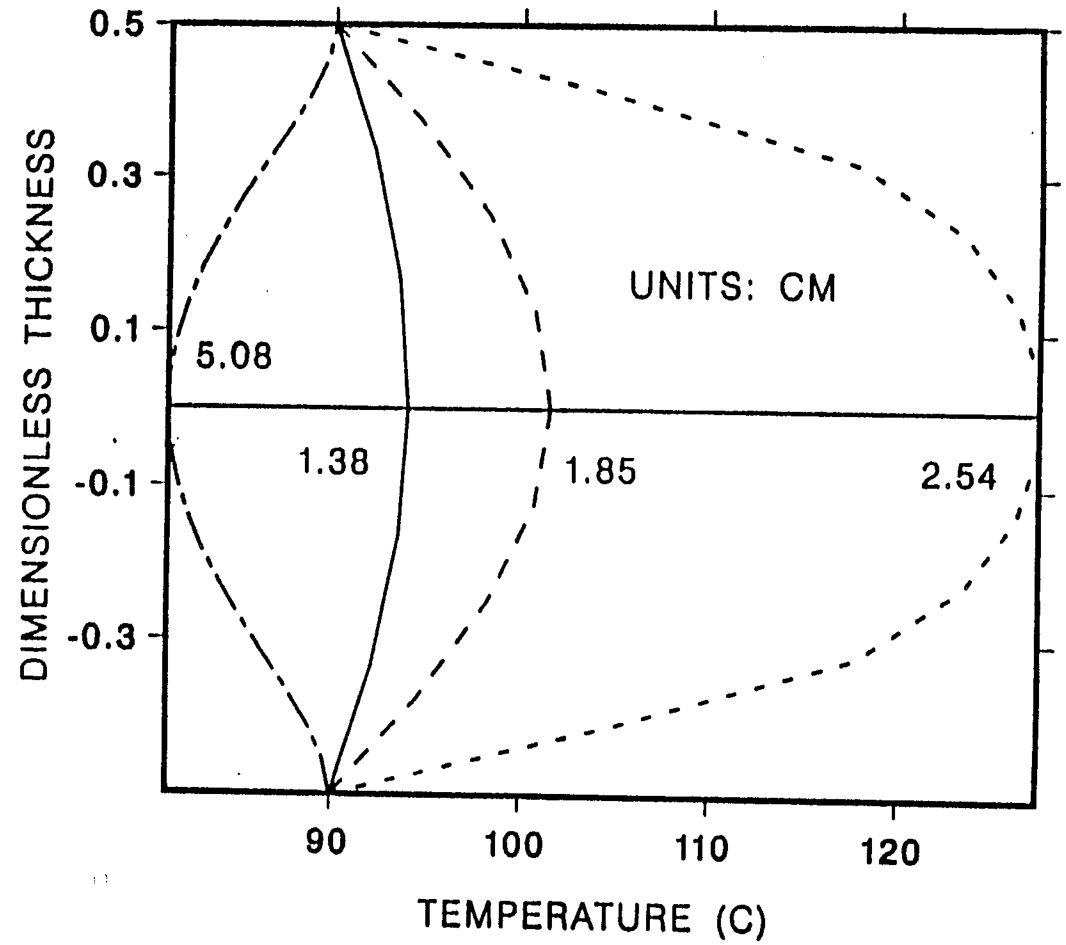

DEGREE OF CURE GRADIENTS AT 164 MINUTES

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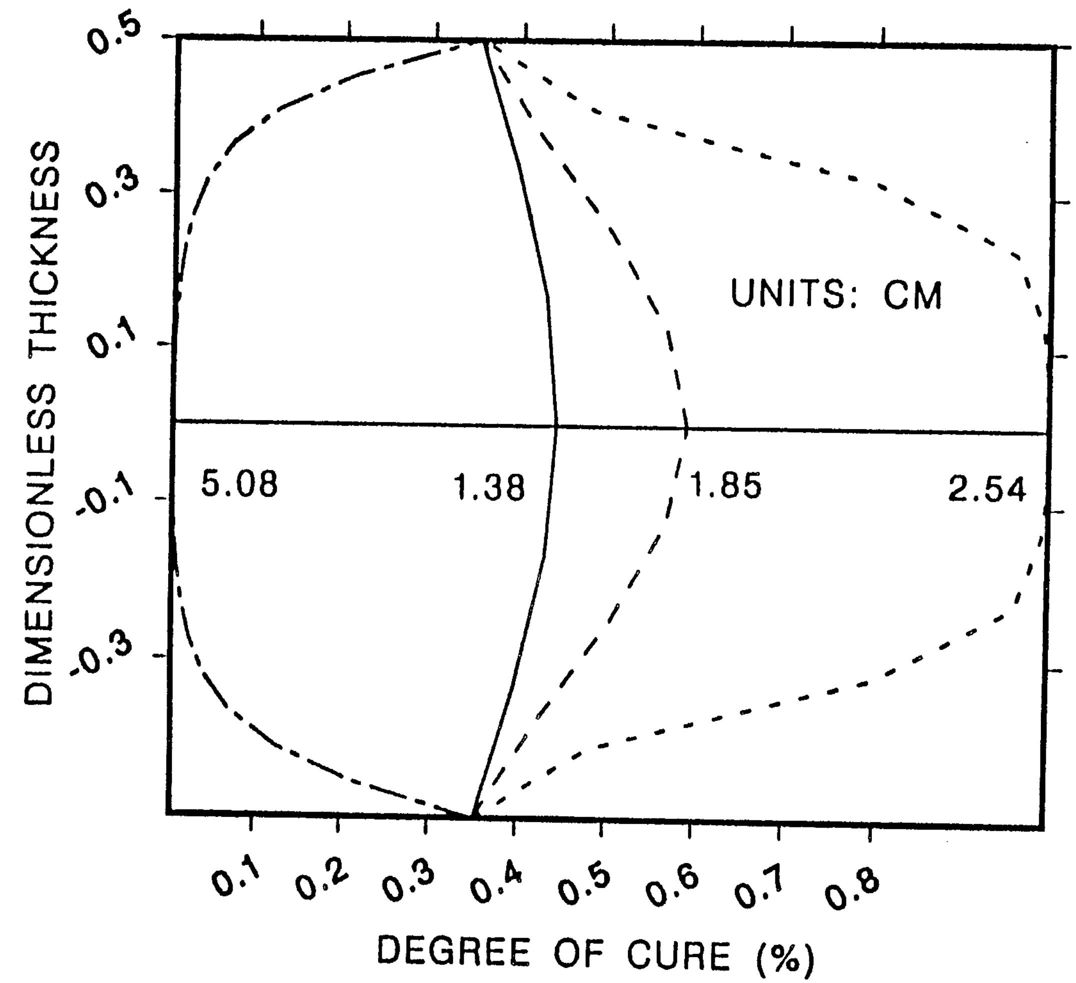

INFLUENCE OF LAMINATE THICKNESS ON CURE STRESS DEVELOPMENT

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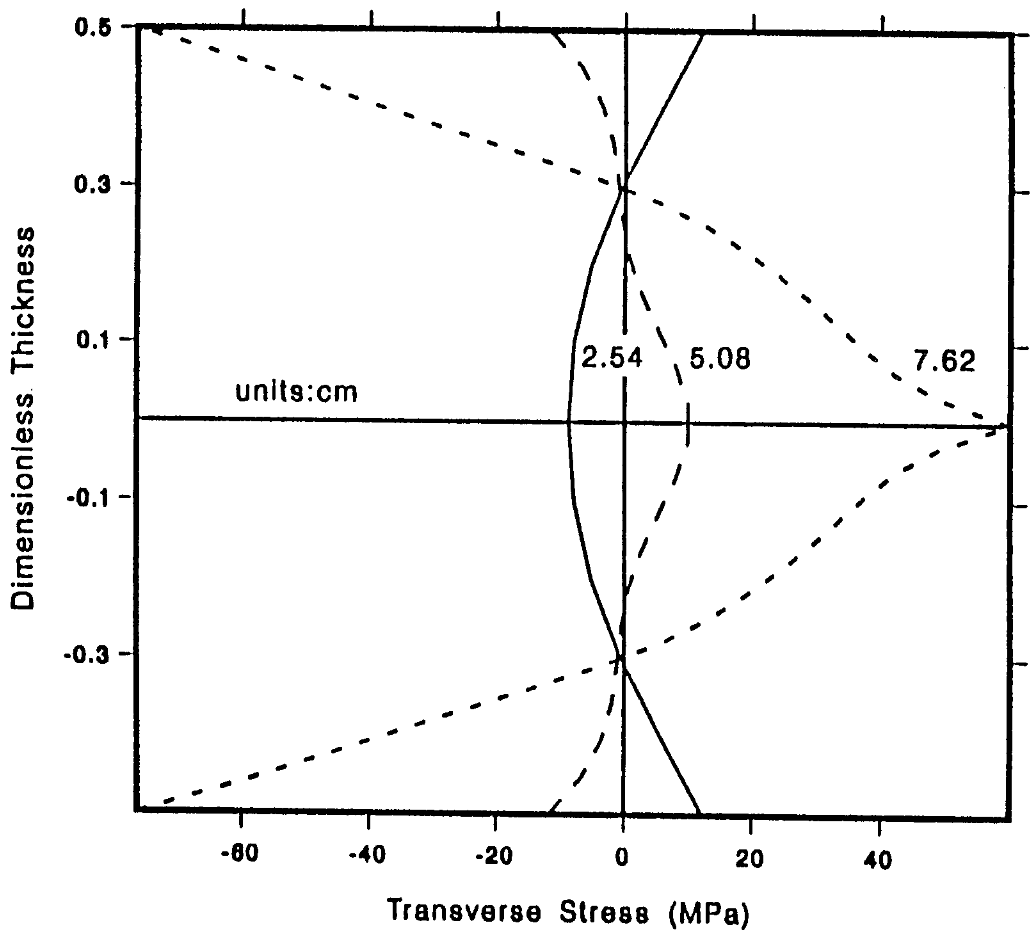

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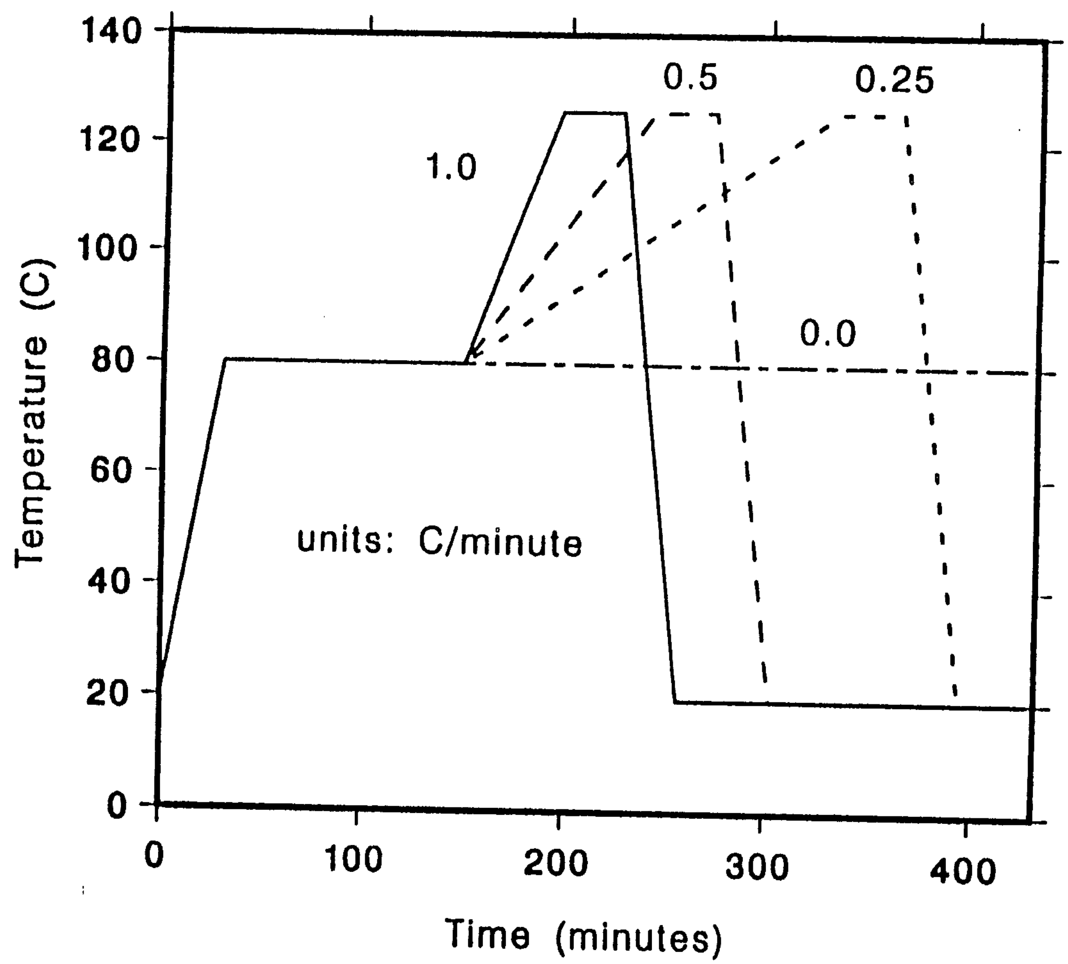
LAMINATE SURFACE TEMPERATURE RAMPS



INFLUENCE OF SURFACE TEMPERATURE RAMP ON CURE STRESS DEVELOPMENT

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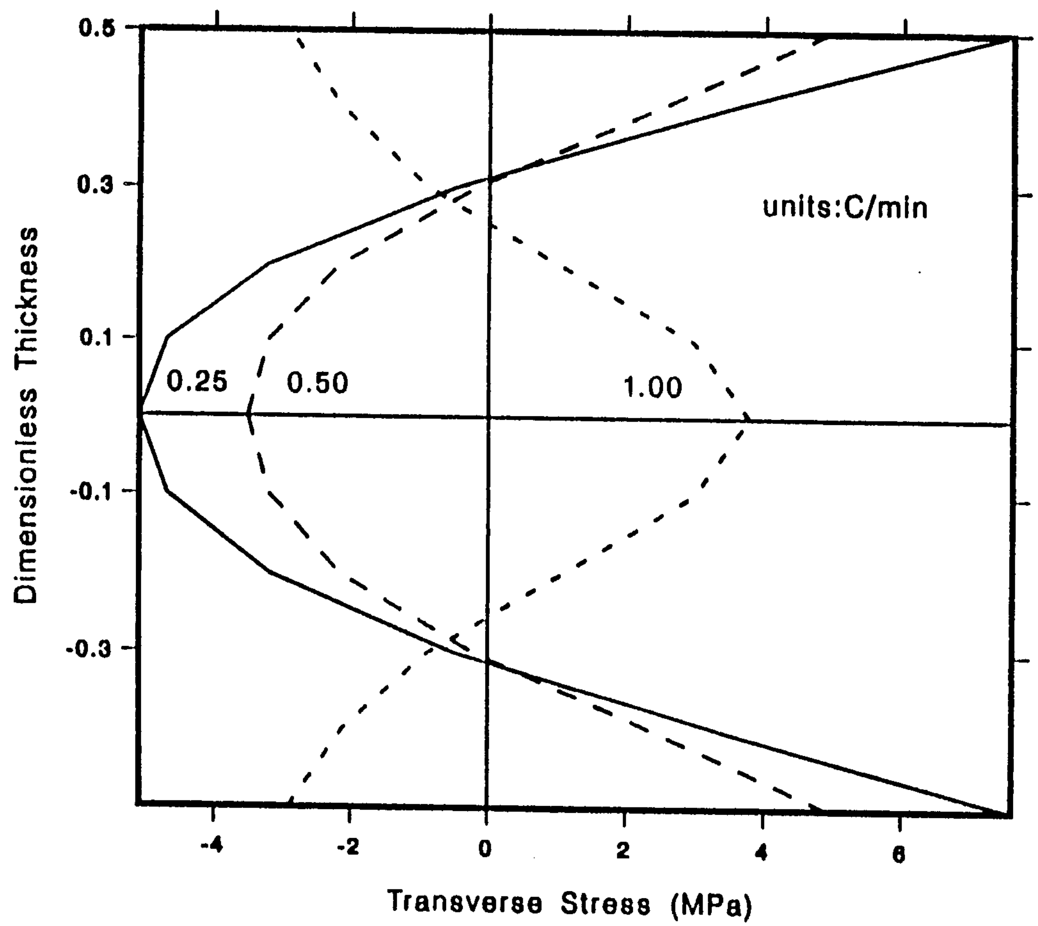

INFLUENCE OF UNSYMMETRIC PROCESSING HISTORY ON CURE STRESS DEVELOPMENT

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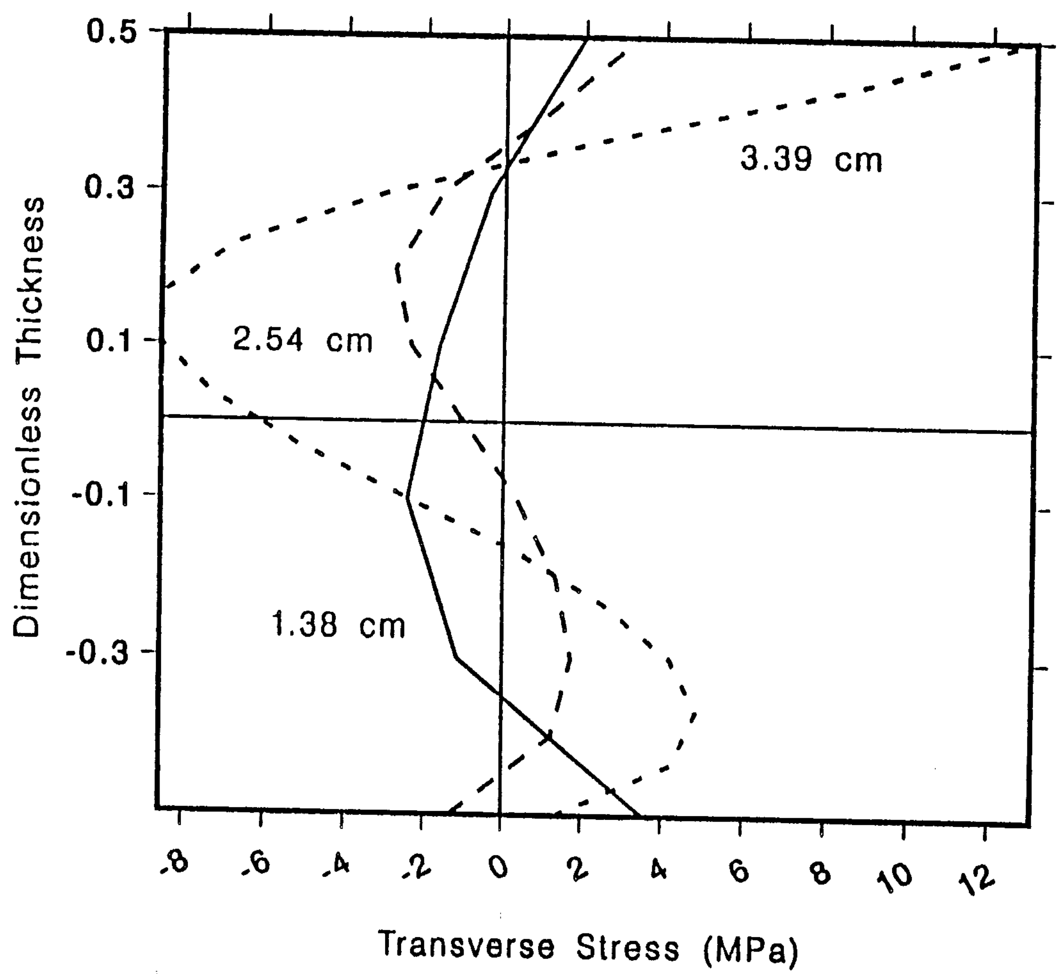

CURE STRESSES SUPERIMPOSED ON STACKING SEQUENCE EFFECTS

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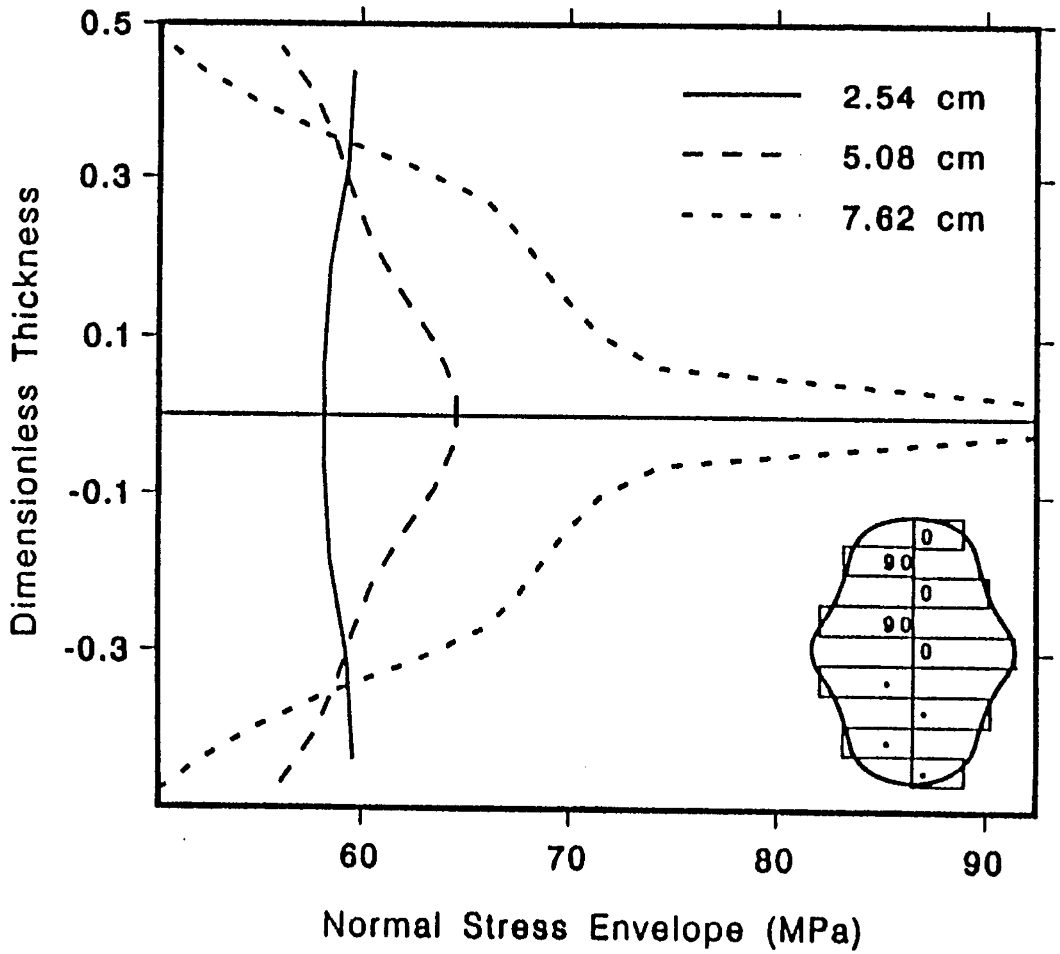

EFFECT OF RESIN CHEMICAL SHRINKAGE ON CURE STRESS DEVELOPMENT

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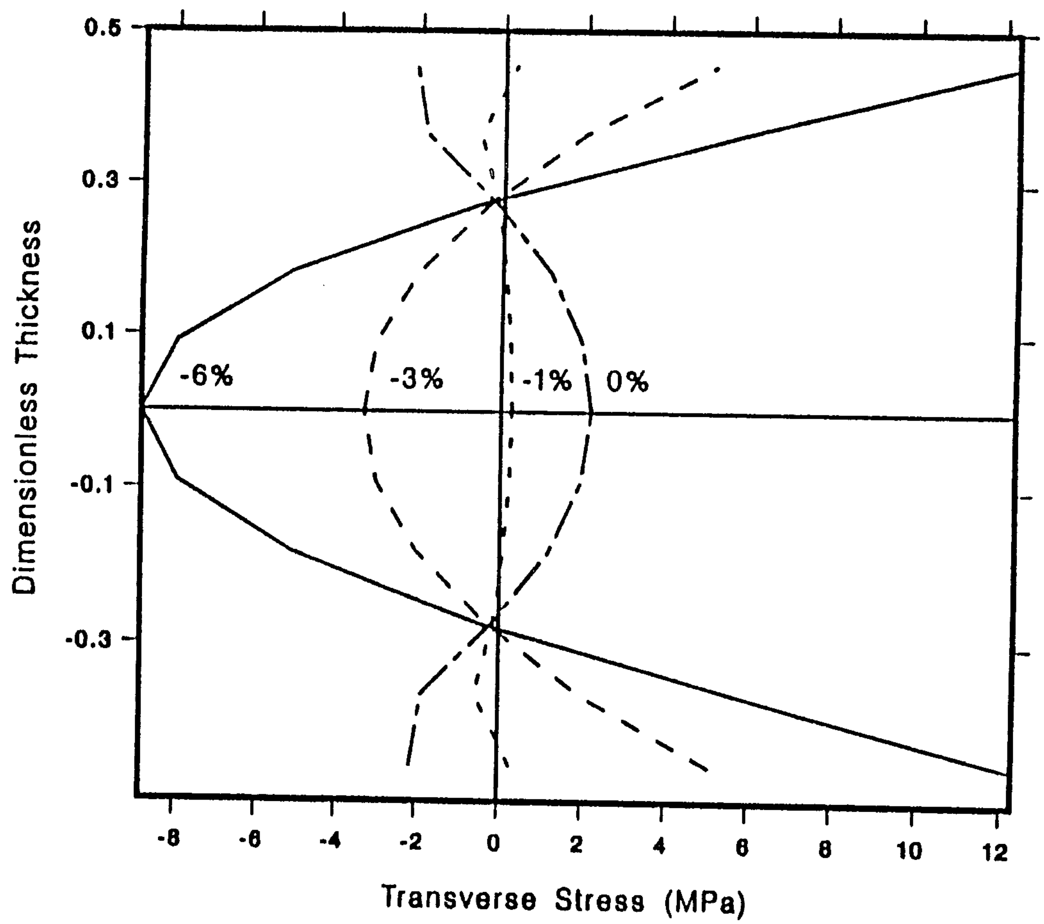

CONCLUSIONS

Thick-section thermosets are susceptible to severe temperature and degree of cure gradients during processing. Many factors influence the processing gradients that develop in thick sections.

- thickness
- thermal properties
- kinetic parameters
- cure cycle
- thermal anisotropy
- part geometry

Gradients in temperature and degree of cure have the potential to induce significant macroscopic stresses in thick sections. Many factors influence process-induced stress and deformation in thick sections.

- laminate thickness
- cure cycle
- stacking sequence
- chemical shrinkage

The mechanics and performance of thick-section thermosetting composites is strongly dependent on the processing history.

CURRENT ACTIVITIES AND FUTURE WORK

- Correlate with Experimental Observations
- Extend to Cylindrical Structures
- Material Model Enhancement
- Cure Cycle Optimization to Enhance Performance
- Couple Model to Failure Theories to Predict Matrix Cracking During Cure
- Apply Methodology to Fiber-Optic Sensor Technology

ULTRASONIC CHARACTERIZATION OF ELASTIC PROPERTIES AND POROSITY OF COMPOSITE MATERIALS

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Robert R. McCormick School of Engineering and Applied Science
Northwestern University
Evanston, IL 60208

ABSTRACT

The anisotropy of fibrous composite materials makes it difficult to determine their elastic constants nondestructively. Quantitative ultrasonic methods were studied to determine elastic constants of a transversely isotropic laminate. Comprehensive relationships of the measured ultrasonic data and material properties were developed. The mode conversion phenomenon at the specimen-liquid interface with oblique incidence ultrasonic immersion testing provided all the necessary information. Determination of elastic constants was possible by measuring wavespeeds and critical angles. Phase velocities were determined by the multiple reflection technique utilizing ultrasonic spectroscopy. A special scanning technique was developed to determine the phase velocities and wave propagation angles in the anisotropic plane from the measured group velocities. The method developed was applied to the characterization of a unidirectional graphite/epoxy composite.

Similar ultrasonic techniques were developed and used for quantitative characterization of porosity introduced during fabrication of the material. This porosity takes the form of dispersed or discrete elongated voids usually at the fiber-matrix interface. During subsequent loading, these voids act as nuclei of further damage growth resulting in strength degradation. Hence, it is important to determine the void content after fabrication.

Since ultrasonic waves are scattered by these voids, the void volume ratio can be directly correlated with material attenuation. However, the measurement of the attenuation coefficient for a thick composite is difficult because of high signal loss.

Special techniques utilizing a pair of transducers in a combination of pulse-echo and through-transmission modes were successfully developed. Methods for determination of material attenuation were discussed in two categories: direct or absolute methods for materials with low signal loss; and indirect or relative methods for materials with higher signal loss. In all cases, transfer functions of the transducers and specimen surfaces were taken into consideration so that the measurement system is self-calibrated.

OBJECTIVES

1. Develop methodology for ultrasonic determination of elastic constants of unidirectional, transversely isotropic composite materials.

2. Develop techniques for measurement of ultrasonic attenuation in composite materials and quantitative correlation with porosity.

WAVE PROPAGATION IN ORTHOTROPIC MATERIAL

Stress-Strain Relations:

$$\begin{Bmatrix}\sigma_{11}\\ \sigma_{22}\\ \sigma_{33}\\ \sigma_{23}\\ \sigma_{31}\\ \sigma_{12}\end{Bmatrix} = \begin{pmatrix} Q_{11} & Q_{12} & Q_{13} & 0 & 0 & 0\\ Q_{12} & Q_{22} & Q_{23} & 0 & 0 & 0\\ Q_{13} & Q_{23} & Q_{33} & 0 & 0 & 0\\ 0 & 0 & 0 & Q_{44} & 0 & 0\\ 0 & 0 & 0 & 0 & Q_{55} & 0\\ 0 & 0 & 0 & 0 & 0 & Q_{66}\end{pmatrix}\begin{Bmatrix}\varepsilon_{11}\\ \varepsilon_{22}\\ \varepsilon_{33}\\ \gamma_{23}\\ \gamma_{31}\\ \gamma_{12}\end{Bmatrix}$$

where

$[Q_{ij}]$ = stiffness matrix
1,2,3 = principal material directions

Then, from wave equation

$$\det\begin{vmatrix}\Gamma_{11}-\rho c^2 & \Gamma_{12} & \Gamma_{13}\\ \Gamma_{12} & \Gamma_{22}-\rho c^2 & \Gamma_{23}\\ \Gamma_{13} & \Gamma_{23} & \Gamma_{33}-\rho c^2\end{vmatrix} = 0$$

where Γ_{ij} are the Christoffel stiffnesses:

$$\Gamma_{11} = n_1^2 Q_{11} + n_2^2 Q_{66} + n_3^2 Q_{55}$$
$$\Gamma_{22} = n_1^2 Q_{66} + n_2^2 Q_{22} + n_3^2 Q_{44}$$
$$\Gamma_{33} = n_1^2 Q_{55} + n_2^2 Q_{44} + n_3^2 Q_{33}$$
$$\Gamma_{12} = n_1 n_2 (Q_{12} + Q_{66})$$
$$\Gamma_{23} = n_2 n_3 (Q_{23} + Q_{44})$$
$$\Gamma_{13} = n_1 n_3 (Q_{13} + Q_{55})$$

RELATIONSHIPS BETWEEN STIFFNESSES AND ENGINEERING CONSTANTS

For orthotropic material with transverse isotropy

$$Q_{12} = Q_{13}$$
$$Q_{33} = Q_{22}$$
$$Q_{44} = \frac{Q_{22} - Q_{23}}{2}$$
$$Q_{66} = Q_{55}$$

Then,

$$E_1 = \frac{Q}{Q_{22}^2 - Q_{23}^2} \qquad E_2 = \frac{Q}{Q_{11}Q_{22} - Q_{12}^2}$$

$$G_{12} = Q_{66} \qquad G_{23} = Q_{44}$$

$$\nu_{12} = \frac{Q_{12}}{Q_{22} + Q_{23}} \qquad \nu_{23} = \frac{Q_{11}Q_{23} - Q_{12}^2}{Q_{11}Q_{22} - Q_{12}^2}$$

where

$$Q = (Q_{22} - Q_{23})\,[Q_{11}\,(Q_{22} + Q_{23}) - 2Q_{12}^2\,]$$

DETERMINATION OF FIVE STIFFNESS CONSTANTS

For wave motion in x_3-direction

$$Q_{22} = Q_{33} = \rho c_{33L}^2$$

where

ρ = material density
c_{33L} = longitudinal wavespeed in x_3-direction

From oblique incidence measurements:

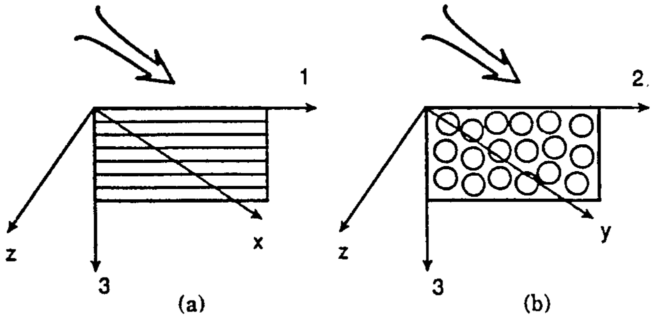
$$Q_{11} = \frac{\rho c_0^2}{\sin^2\theta_{cr}^L}$$

$$Q_{55} = Q_{66} = \frac{\rho c_0^2}{\sin^2\theta_{cr}^T}$$

where

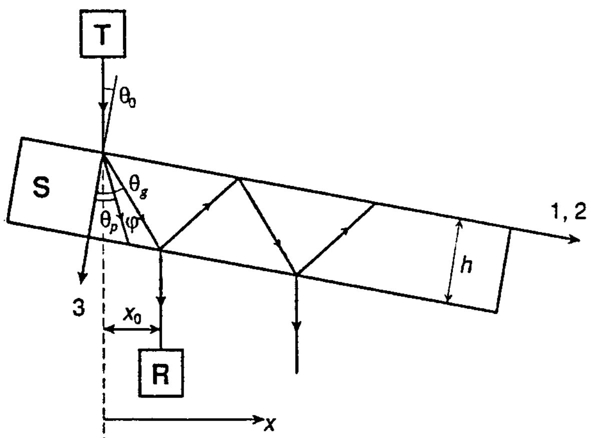
c_0 = wavespeed in water
$\theta_{cr}^L, \theta_{cr}^T$ = critical angles for quasi-longitudinal and quasi-transverse waves, respectively.

Constants Q_{13} and Q_{23} are obtained from determinantal wave equation.

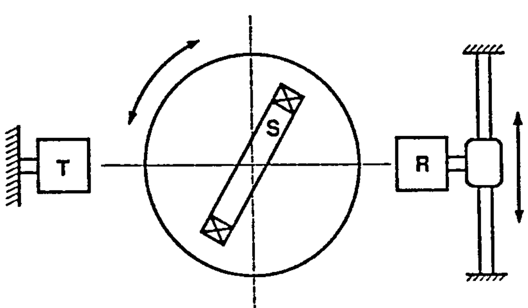


Schematic diagrams of oblique incidence immersion testing for determination of composite stiffnesses.

WAVE PROPAGATION AND ENERGY FLUX



θ_0 = angle of incidence
θ_p = propagation angle of wavefront
θ_g = propagation angle of energy flux
φ = $\theta_g - \theta_p$



Schematic of experimental setup. S = Specimen, T = Transmitting transducer, R = Receiving transducer.

DETERMINATION OF ENERGY FLUX VECTOR

Transmitted energy

$$P = \int_{-\infty}^{\infty} |s(t)|\, dt$$

where $s(t)$ = received waveform

Offset of received ray

$$x_0 = \frac{\int Px\, dx}{\int P\, dx}$$

Propagation direction of energy flux

$$\tan\theta_g = \frac{x_0}{h \cos\theta_0} + \tan\theta_0$$

where

h = specimen thickness
θ_0 = angle of incidence

DETERMINATION OF WAVE PROPAGATION DIRECTION

Group velocity c_g:

$$\frac{1}{c_g} = \frac{\Delta t \cos\theta_g}{2h} + \frac{\sin\theta_g \sin\theta_0}{c_0}$$

where Δt = time interval between successive echoes

Phase velocity c_p:

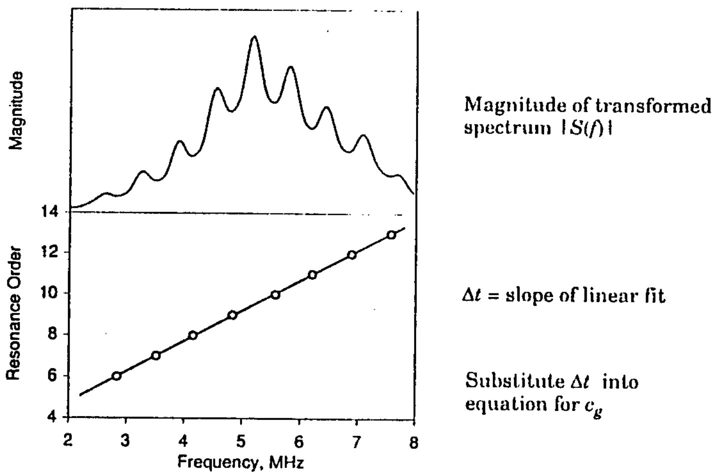
$$c_p = c_g \cos\varphi = c_g \cos(\theta_g - \theta_p)$$

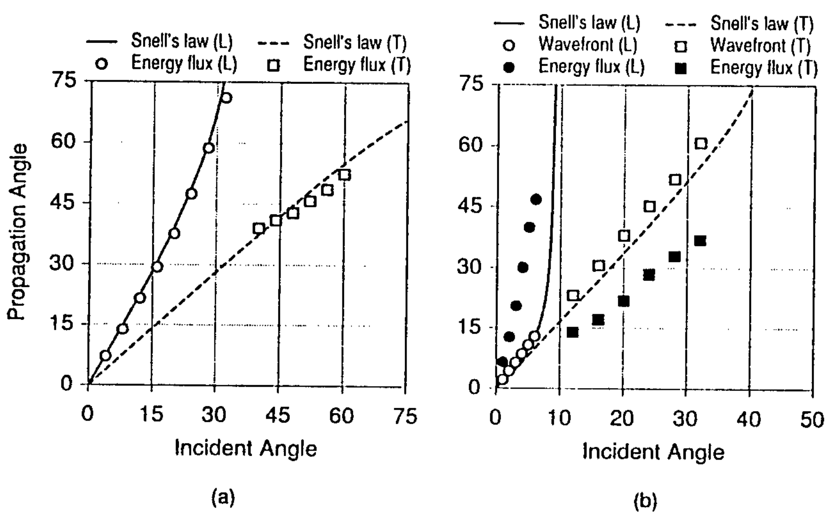
Direction of wave propagation:

$$\tan\theta_p = \frac{c_g \sin\theta_0 \cos\theta_g}{c_0 - c_g \sin\theta_0 \sin\theta_g}$$

TECHNIQUE FOR MEASUREMENT OF GROUP VELOCITY

Waveform in frequency domain: $S(f) = \int_{-\infty}^{\infty} s(t) \exp(-j2\pi ft)\, dt$

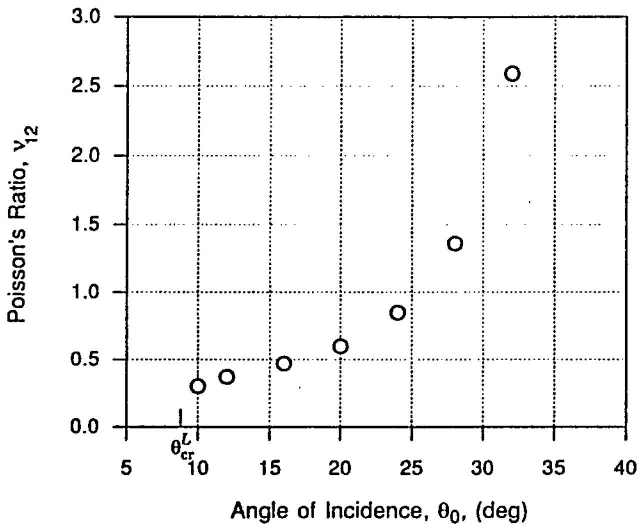




Propagation angles of wavefront and energy flux for (a) Plexiglas and (b) graphite/epoxy samples. Letters in parentheses denote quasi-longitudinal (L) and quasi-transverse (T) wave modes.

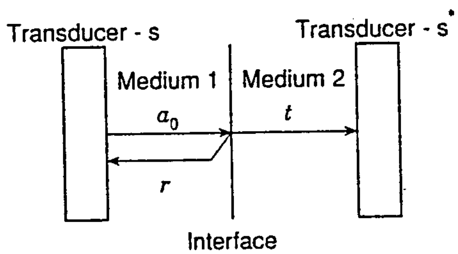
TABLE 1. Measured elastic properties of unidirectional graphite/epoxy laminate.

Material property	Experimental measurements	
	Mechanical	Ultrasonic
Longitudinal modulus, E_1	145 GPa (21.00 Msi)	140 GPa (20.30 Msi)
Transverse modulus, E_2	10.6 GPa (1.54 Msi)	11.3 GPa (1.64 Msi)
Shear modulus, G_{12}	7.6 GPa (1.10 Msi)	7.4 GPa (1.07 Msi)
Shear modulus, G_{23}	3.9 GPa (0.56 Msi)	3.8 GPa (0.56 Msi)
Poisson's ratio, ν_{12}	0.27	0.30
Poisson's ratio, ν_{23}	0.50	0.49



Poisson's ratio as determined for various angles of incidence.

SYSTEM RESPONSE



Time Domain Waveforms

$$r(\tau) = a_0(\tau) * q_1(\tau) * r_{12}(\tau) * s(\tau) * q_1(\tau)$$
$$t(\tau) = a_0(\tau) * q_1(\tau) * t_{12}(\tau) * q_2(\tau) * s^*(\tau)$$

where q_1, q_2 = material response
r_{12} = reflection coefficient
t_{12} = transmission coefficient
s, s^* = transducer sensitivities

Frequency Domain Representation

$$R(\omega) = A_0(\omega)\, Q_1^2(\omega)\, R_{12}(\omega)\, S(\omega)$$
$$T(\omega) = A_0(\omega)\, Q_1(\omega)\, T_{12}(\omega)\, Q_2(\omega)\, S^*(\omega)$$

where $F(\omega) = \int_{-\infty}^{\infty} f(\tau) \exp(-j\omega\tau)\, d\tau$

SUMMARY AND CONCLUSIONS

- Determination of elastic constants was possible by measuring wavespeeds and critical angles under normal and oblique incidence.
- Energy flux and wave propagation directions and phase velocities were accurately measured by determining the ray offset (x_0) in through-transmission mode.
- Properties measured by mechanical and ultrasonic methods, in general, are in good agreement.
- Out-of-plane properties were measured accurately and more easily by the ultrasonic method than by mechanical testing.
- Limitations: The material must be unidirectional and transversely isotropic. Poisson's ratio ν_{12} cannot be measured accurately at this point.
- The sensitivity of Poisson's ratio ν_{12} determination to angle of incidence is high and it reaches its lowest value at the critical angle θ_{cr}^{L}.

DETERMINATION OF MATERIAL ATTENUATION

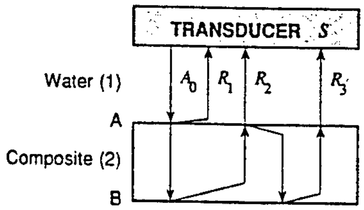
DIRECT (ABSOLUTE) METHODS

- Multiple Echo
- Combined Pulse Echo and Through-Transmission

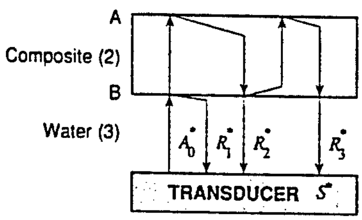
INDIRECT (RELATIVE) METHODS

- Single Through-Transmission
- Double Through-Transmission
- Combined Pulse Echo and Through-Transmission

MULTIPLE ECHO TECHNIQUE



$$R_1 = S\,A_0\,Q_1^2\,R_{12}$$
$$R_2 = S\,A_0\,Q_1^2\,Q_c^2\,T_{12}\,T_{21}\,R_{23}$$
$$R_3 = S\,A_0\,Q_1^2\,Q_c^4\,T_{12}\,T_{21}\,R_{23}^2\,R_{21}$$



$$R_1^* = S^* A_0^* Q_3^2\,R_{32}$$
$$R_2^* = S^* A_0^* Q_3^2\,Q_c^2\,T_{32}\,T_{23}\,R_{21}$$
$$R_3^* = S^* A_0^* Q_3^2\,Q_c^4\,T_{32}\,T_{23}\,R_{21}^2\,R_{23}$$

Reflection coefficients

$$R_A = |R_{12}^A| = |R_{21}^A|$$
$$R_B = |R_{12}^B| = |R_{21}^B|$$

Transmission coefficients

$$T_{12}\,T_{21} = 1 - R_A^2$$
$$T_{23}\,T_{32} = 1 - R_B^2$$

MULTIPLE ECHO TECHNIQUE - cont.

Amplitude Ratios

$$\frac{R_2}{R_1} = Q_c^2\,[1 - R_A^2]\,\frac{R_B}{R_A}$$
$$\frac{R_3}{R_2} = \frac{R_3^*}{R_2^*} = Q_c^2\,R_B\,R_A$$
$$\frac{R_2^*}{R_1^*} = Q_c^2\,[1 - R_B^2]\,\frac{R_A}{R_B}$$

Then,

$$Q_c^2 = \frac{R_3}{R_2}\left(\frac{R_1 R_3 + R_2^2}{R_1 R_3}\,\frac{R_1^* R_3^* + R_2^{*2}}{R_1^* R_3^*}\right)^{1/2}$$

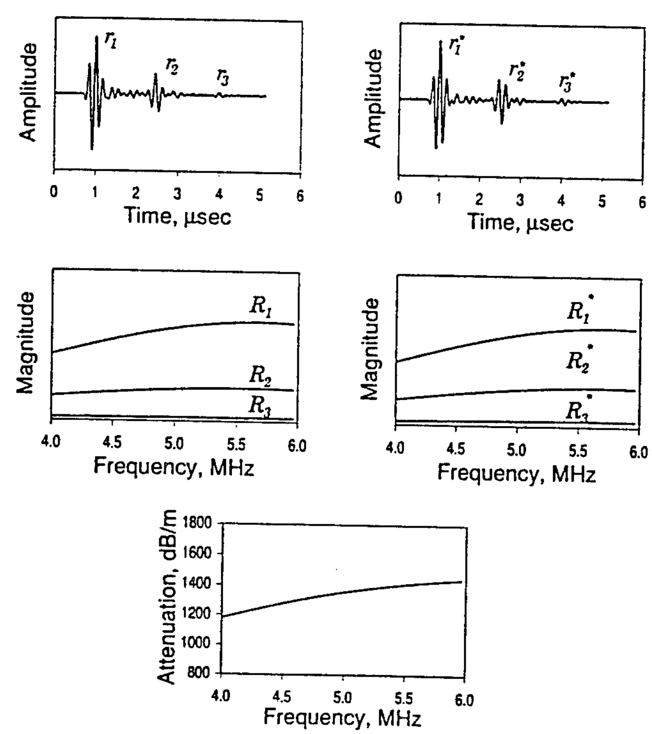
where Q_c = amplitude attenuation in composite specimen

Material Attenuation Coefficient

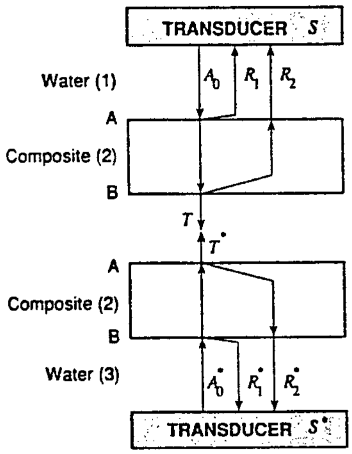
$$\alpha = -\frac{20}{h}\log Q_c \qquad \text{(dB / unit length)}$$

MULTIPLE ECHO TECHNIQUE - cont.

Typical results at point of 3.8% porosity



COMBINED PULSE ECHO AND THROUGH-TRANSMISSION TECHNIQUE



$$R_1 = S A_0 Q_1^2 R_{12}$$
$$R_2 = S A_0 Q_1^2 Q_c^2 T_{12} T_{21} R_{23}$$
$$T = S^* A_0 Q_1 Q_3 Q_c T_{12} T_{23}$$

$$R_1^* = S^* A_0^* Q_3^2 R_{32}$$
$$R_2^* = S^* A_0^* Q_3^2 Q_c^2 T_{32} T_{23} R_{21}$$
$$T^* = S A_0^* Q_3 Q_1 Q_c T_{32} T_{21}$$

COMBINED PULSE ECHO AND THROUGH-TRANSMISSION TECHNIQUE - cont.

Amplitude Ratios

$$\frac{R_2}{T} = \frac{S}{S^*}\frac{Q_1}{Q_3} Q_c \frac{T_{21}}{T_{23}} R_B$$
$$\frac{R_2}{R_1} = \frac{R_2^*}{R_1^*} = Q_c^2 [1 - R_A^2] \frac{R_B}{R_A}$$
$$\frac{R_2^*}{T^*} = \frac{S^*}{S}\frac{Q_3}{Q_1} Q_c \frac{T_{23}}{T_{21}} R_A$$

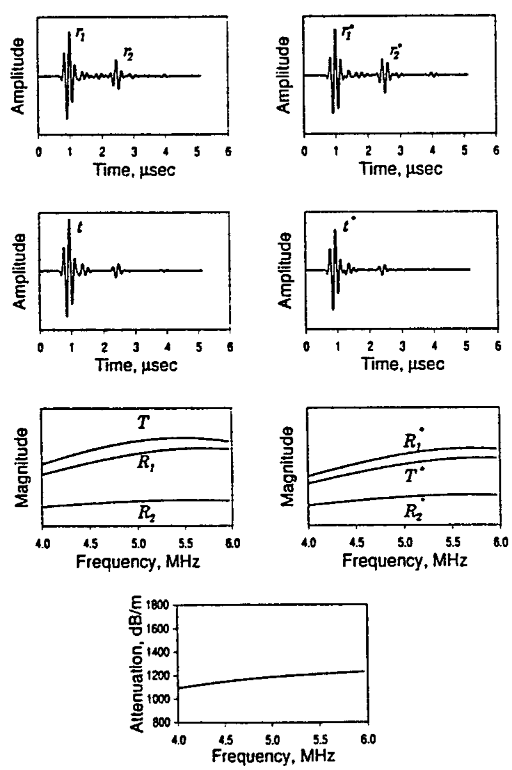
Then

$$Q_c^2 = \frac{R_2 R_2^*}{TT^*}\left(\frac{R_1 R_2^* + TT^*}{R_1 R_2^*}\right)^{1/2}\left(\frac{R_1^* R_2 + TT^*}{R_1^* R_2}\right)^{1/2}$$

Material Attenuation Coefficient

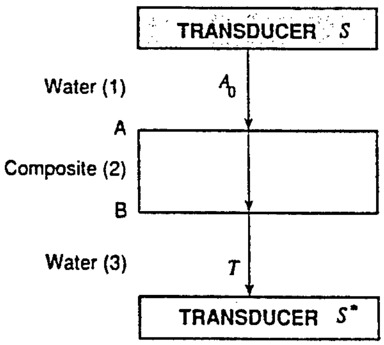
$$\alpha = -\frac{20}{h}\log Q_c$$

COMBINED PULSE ECHO AND THROUGH-TRANSMISSION TECHNIQUE - cont.



INDIRECT (RELATIVE) METHODS

SINGLE THROUGH-TRANSMISSION



$$T = S^* A_0 \, Q_1 \, Q_3 \, Q_c \, T_{12} T_{23}$$
$$T_{ref} = S^* A_0 \, Q_1 \, Q_3 \, (Q_c)_{ref} \, T_{12} \, T_{23}$$

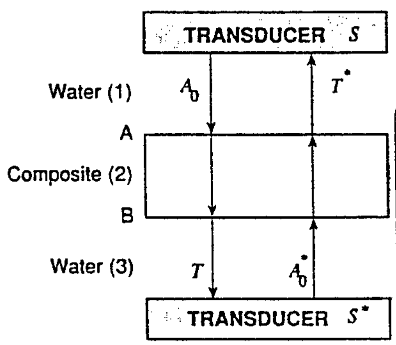
Relative Attenuation

$$\eta = \frac{Q_c}{(Q_c)_{ref}} = \frac{T}{T_{ref}}$$

Absolute Material Attenuation

$$\alpha = \alpha_{ref} - \frac{20}{h} \log \eta$$

DOUBLE THROUGH-TRANSMISSION



$$T = S^* A_0 \, Q_1 \, Q_3 \, Q_c \, T_{12} \, T_{23}$$
$$T^* = S \, A_0^* \, Q_1 \, Q_3 \, Q_c \, T_{32} \, T_{21}$$
$$T_{ref} = S^* A_0 \, Q_1 \, Q_3 \, (Q_c)_{ref} \, T_{12} \, T_{23}$$
$$T_{ref}^* = S \, A_0^* \, Q_1 \, Q_3 \, (Q_c)_{ref} \, T_{32} \, T_{21}$$

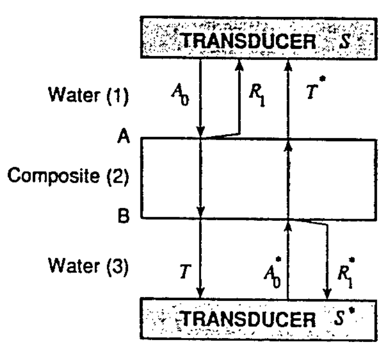
Relative Attenuation

$$\eta = \frac{Q_c}{(Q_c)_{ref}} = \left(\frac{T \, T^*}{(T \, T^*)_{ref}} \right)^{1/2}$$

Absolute Material Attenuation

$$\alpha = \alpha_{ref} - \frac{20}{h} \log \eta$$

COMBINED REFLECTION AND TRANSMISSION TECHNIQUE



$$R_1 = S \, A_0 \, Q_1^2 \, R_{12}$$
$$R_1^* = S^* A_0^* \, Q_3^2 \, R_{32}$$
$$T = S^* A_0 \, Q_1 \, Q_3 \, Q_c \, T_{12} T_{23}$$
$$T^* = S A_0^* \, Q_3 \, Q_1 \, Q_c \, T_{32} \, T_{21}$$

Relative Attenuation

$$\frac{T \, T^*}{R_1 \, R_1^*} = \frac{T_{12} \, T_{21} \, T_{32} \, T_{23}}{R_{12} \, R_{32}} \, Q_c^2 = K \, Q_c^2$$

$$\eta = \frac{Q_c}{(Q_c)_{ref}} = \left\{ \frac{T \, T^* / R_1 \, R_1^*}{(T \, T^* / R_1 \, R_1^*)_{ref}} \right\}^{1/2}$$

Absolute Material Attenuation

$$\alpha = \alpha_{ref} - \frac{20}{h} \log \eta$$

POROSITY MEASUREMENT

Thresholding

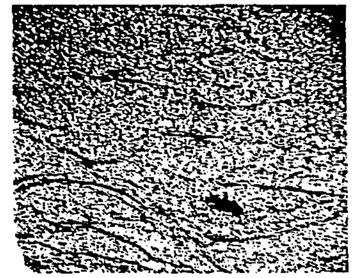
$$g(x,y) = \begin{cases} a & \text{if } f(x,y) < T \\ b & \text{if } f(x,y) \leq T \end{cases}$$

where

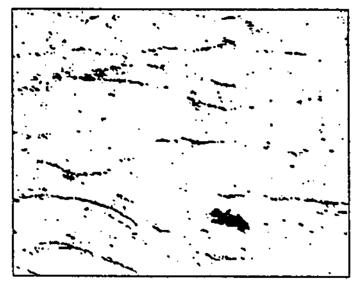
$f(x,y)$	=	gray scale (image) function of photomicrograph of specimen section
$g(x,y)$	=	thresholded image function
T	=	threshold value

Void Volume Ratio

$$V_V = \frac{\text{Number of thresholded pixels}}{\text{Total number of pixels}}$$

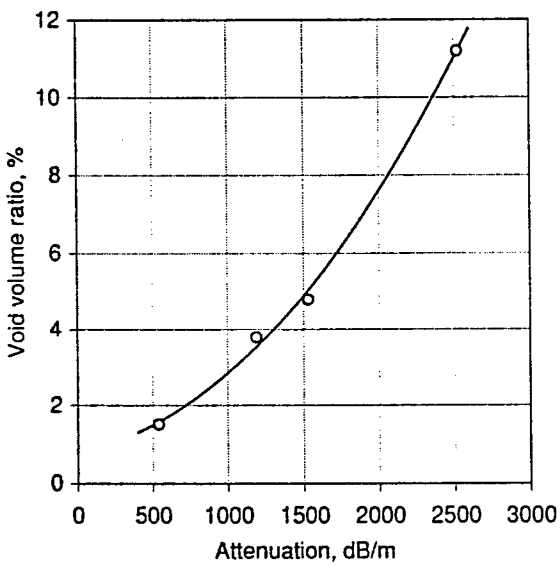


(a)

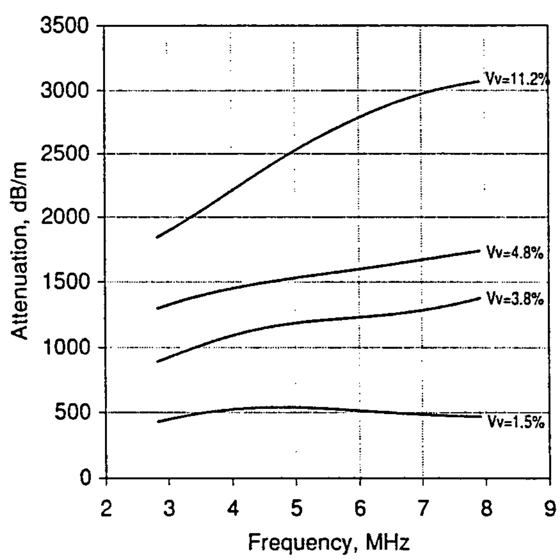


(b)

Photomicrographs of specimen section at point of 3.8% porosity.
(a) as-obtained image, (b) thresholded image



Correlation of porosity and material attenuation at 5 MHz frequency.



Variation of material attenuation with frequency for various values of porosity.

SOME MECHANICS PROBLEMS OF COMPOSITE MATERIALS

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Abstract — Substantial progress has been made on the mechanics of interface fracture. The recent development is assessed in an Acta-Scripta Metallurgica Proceeding edited by Ruhle, Evans, Ashby and Hirth (1990). An engineering program has emerged which allows the fracture resistance of interfaces to be measured and utilized. Specimen geometries suitable for fracture testing are rigorously calibrated (Suo and Hutchinson, 1989; Charalambides et al., 1989; O'Dowd, Shih and Stout, 1990). The program has been implemented in experiments by several groups (Cao and Evans, 1989; Wang and Suo, 1990; Liechti and Chai, 1990; Stout, O'Dowd and Shih, 1990). Some progress (Shih et al., 1990) in the plasticity aspects of bimaterial interfaces is summarized.

1. Delamination of Composite Laminates

A comprehensive framework to assess whether cracks extend along, or kink out of interfaces has been formulated. The framework makes advantageous use of several unifying concepts, the H tensor, eigenvectors $\mathbf{w}$ and $\mathbf{w}_3$, generalized interfacial traction components t and t_3, and mode mixity $\hat{\psi}$ and ϕ. The framework is completely general in that no special material symmetry is assumed. The explicit results tabulated for orthotropic bimaterials allow immediate applications to bicrystals and cross-ply laminates. Several crack geometries, which are suitable for evaluating the fracture resistance of uniaxial and cross-ply composites, are being calibrated (Bao et al., 1990; Choi, Shih and Suo, 1990).

A framework to quantify interface fracture resistance under mixed mode conditions is proposed. The concepts of mode mixity and toughness surface are unified by using generalized interface traction components based on the eigenvectors of an algebraic eigenvalue equation involving H, a 3 by 3 positive definite Hermitian matrix depending on the elastic constants of the two materials and having dimension of compliance (Suo, 1990; Wang, Shih and Suo, 1990)

2. R-Curve Phenomena

Over the last decade, it has become increasingly clear that fracture resistance of brittle materials can be enhanced by a variety of bridging mechanisms. The mechanics language that describes this is resistance curves (R-curves): toughness increase as crack grows. The R-curve behaviors appear in the delamination of unidirectional or laminated composites. In ceramic matrix composites, this is largely due to the intact fibers left behind the crack front, where the crack switches from one fiber-matrix interface to another. For polymer composites additional resistance is provided by matrix damage in the form of voids, craze ot micro-cracks. The three dimensional architechture of fiber threading across prospective delamination planes may give rise to substantial fracture resistance.

3. Toughening of Ceramics by Ductile Layers

Brittle solids can be toughened by incorporating ductile inclusions into them. When these composites are loaded, the plastic deformation in the ductile phase is constrained by the surrounding elastic matrix and high stresses develop. The high stresses trigger various failure mechanisms, which may not operate under unconstrained plastic deformation conditions. A type of failure occurring only under constrained plastic deformation was observed in a recent experiment of Reimanis and Evans (1990). They have prepared a cracked specimen, made of a thin foil of gold diffusion bonded between plates of sapphire The interfacial cracks were introduced by emplacing a hardness indentation in the center of the tensile face of the specimen and then loading it under three point bending. Thus a crack grew unstably from the indentation through the sapphire and reached the interface. The crack then bifurcated along the interface and subsequently arrested. The cracked specimen was subsequently loaded under four point bending in a dry nitrogen environment. They observed interfacial cavities developing ahead of the crack tip at a distance of the order of the foil thickness. The nucleated cavities did not coalesce with the crack tip. Instead with further loading new cavities nucleated at even larger distances from the crack tip. This failure mechanism is attributed to the development of high normal interfacial tractions initially at a distance of the order of the layer thickness ahead of the crack tip and subsequently at even larger distances as a consequence of the constrained plastic deformation of the

gold foil.

4. Elasticity and Plasticity Aspects of Bimaterial Interfaces

In so-called advanced materials such as structural ceramics, ceramic and metal-matrix composites and polycrystalline intermetallic alloys, interfacial and intergranular fractures are common and may, in large part, determine the material's overall mechanical response. For many material systems it is the low fracture toughness that limits their use in engineering and structural components. The need to understand, quantify and improve the toughness of advanced materials has renewed interest in the elastic interface crack problem.

The plasticity aspects of cracks on bimaterial interfaces in which one (or more) of the constituent materials can deform plastically, are beginning to receive attention and some important results have emerged from recent studies (e.g., Shih et al., 1988,1989,1990) The small scale yielding solutions are members of a family parameterized by a near-tip phase angle ξ and the magnitude of the crack tip fields nearly scales with the $J-$integral. There are surprising effects, for example, plastic zones and regions of finite plastic strains that develop at bimaterial interfaces are considerably larger than those in similarly loaded homogeneous bodies. Furthurmore the near-tip triaxail stress states in the weaker material as well as in the stronger material can be higher than those that develop near cracks in a homogeneous material. Our studies provide a basis for phrasing crack growth along the interface in terms of a phase dependent critical J value which we shall designate by $J_C(\xi)$.

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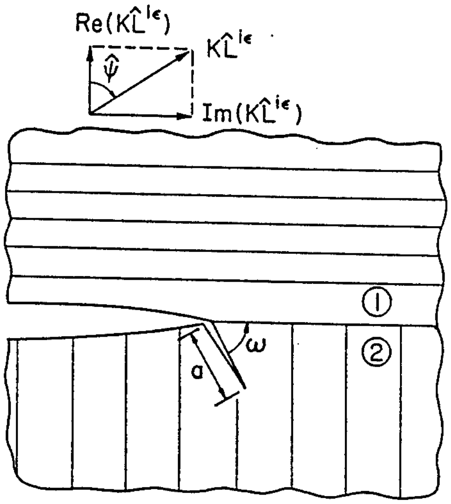
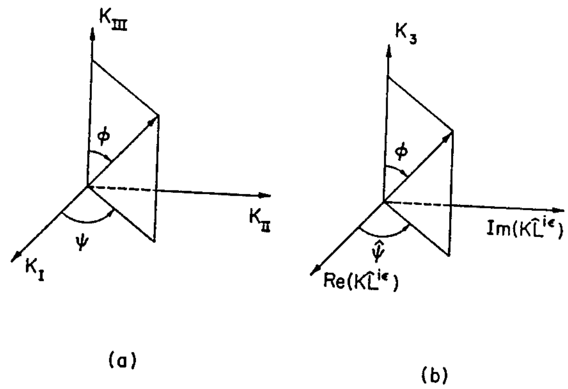
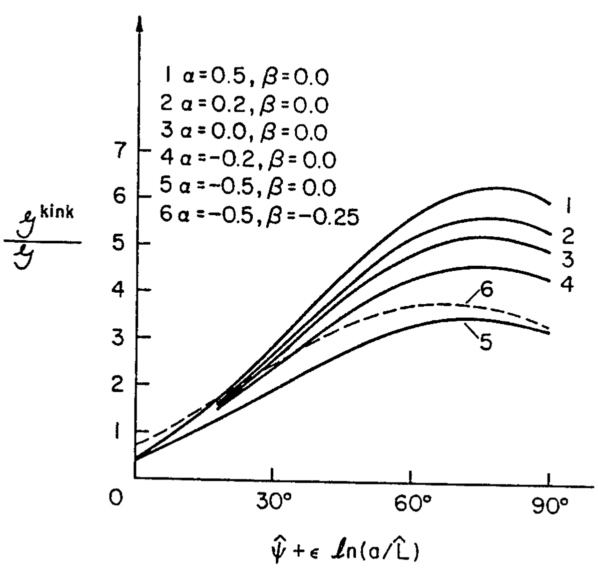
CRACK EXTENSION AND KINKING IN LAMINATES AND BICRYSTALS



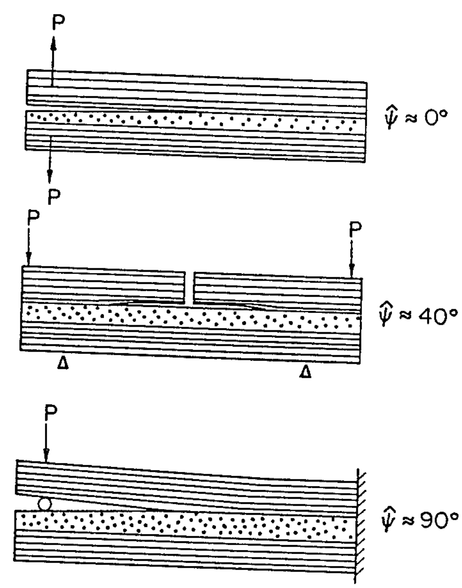
COMPETITION BETWEEN CRACK EXTENSION AND KINKING



MODE MIXITIES DEFINED AS SOLID ANGLES IN K SPACE

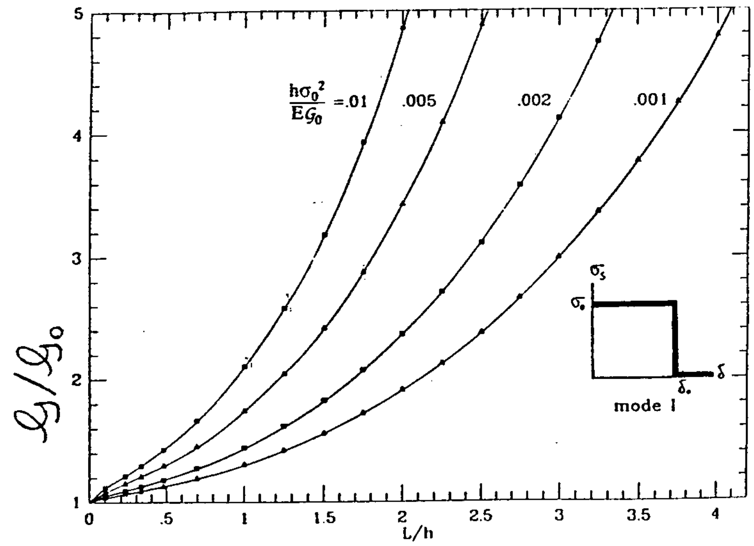


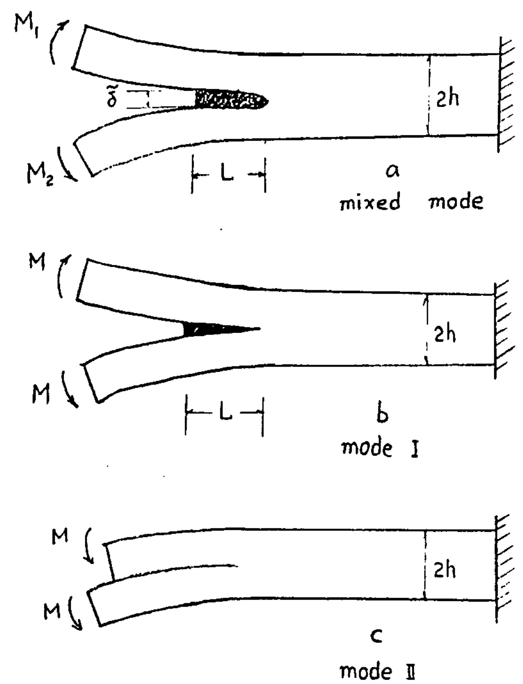
G^{kink}/G AS A FUNCTION OF THE LOADING PHASE

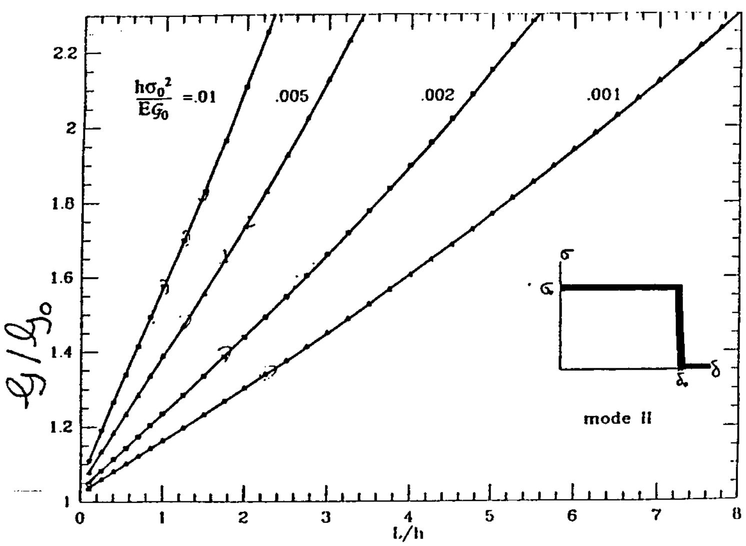


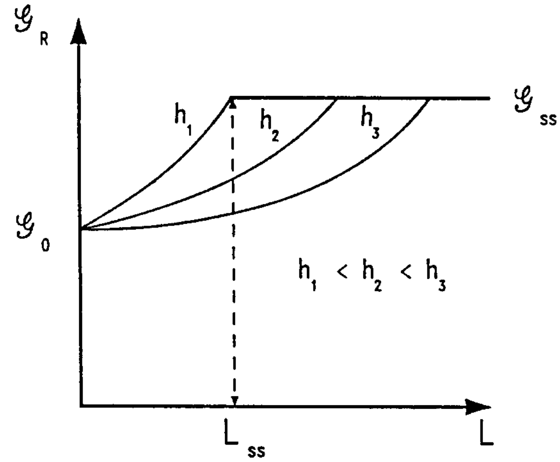
SPECIMENS FOR DETERMINATION OF MIXED MODE DELAMINATION TOUGHNESS FOR CROSS-PLY LAMINATES

DELAMINATION R-CURVE PHENOMENA DUE TO DAMAGE

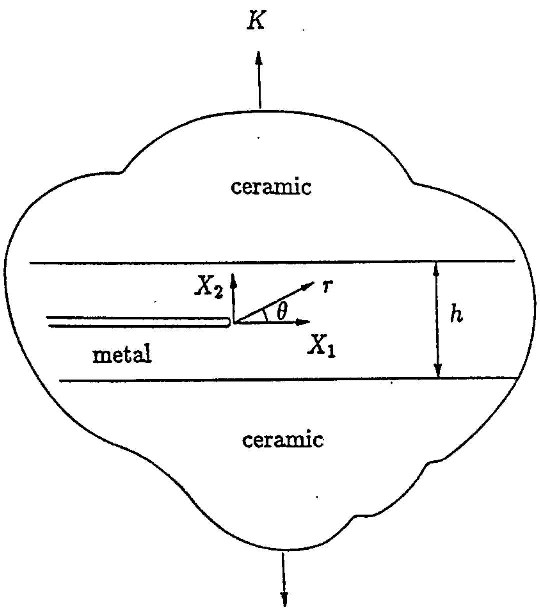






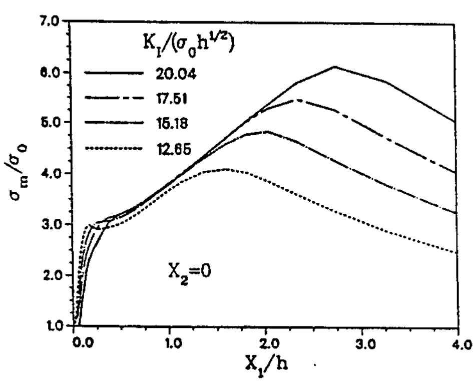


FAILURE MECHANISMS OF CERAMIC/MATRIX COMPOSITES

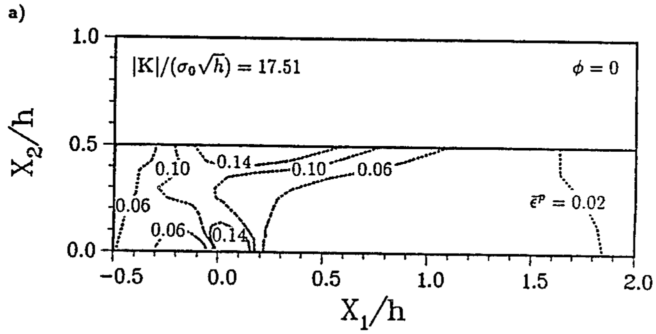


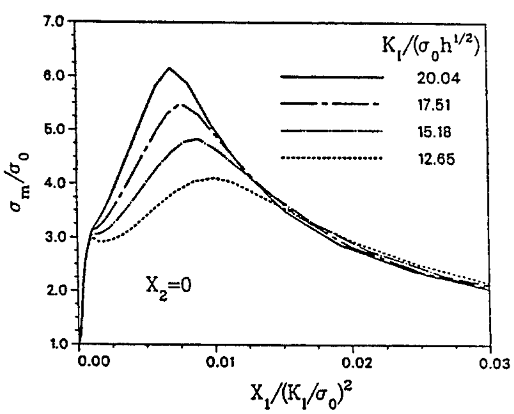
A metal foil is bonded between two ceramic substrates, with a centerline crack. The elastic K-field is applied at distances large compared to the foil thickness.

Under mode I loading the plastic deformation is constrained symmetrically with respect to the crack plane and high stresses develop through the foil thickness. The peak stresses develop at distances of the order of the foil thickness away from the crack tip and increase continuously with remote loading.



a)





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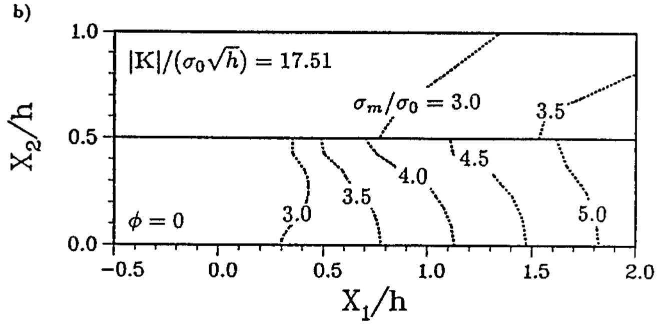
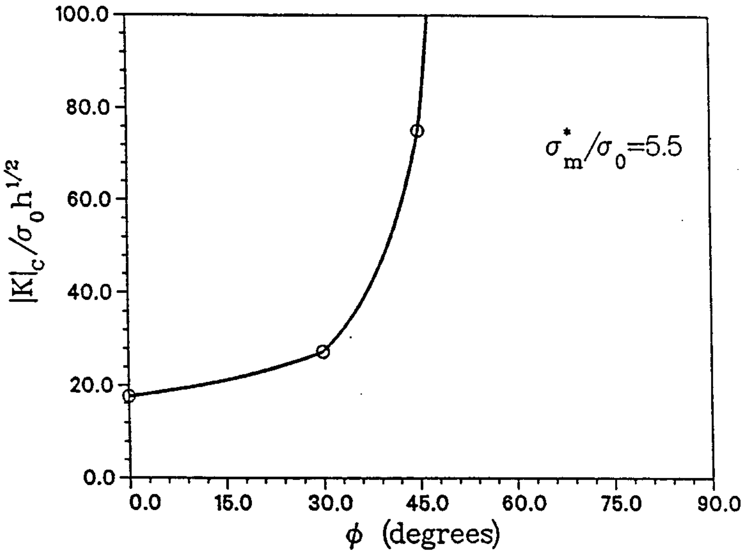
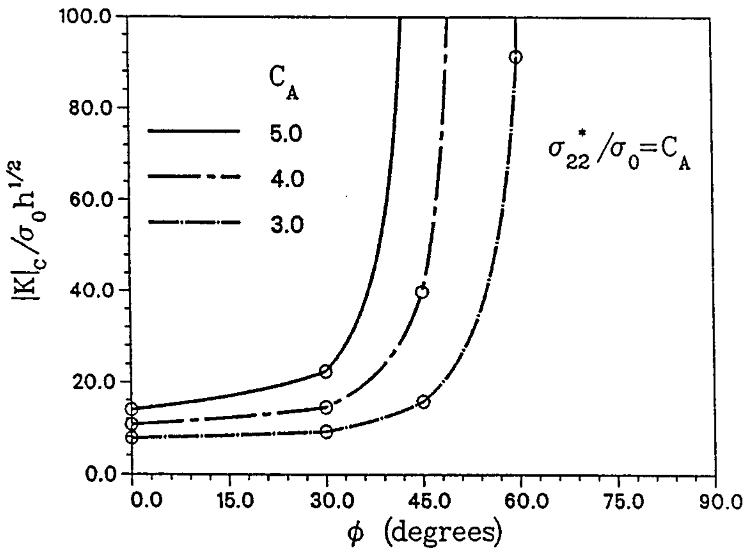


FAILURE MECHANISMS OF CERAMIC/MATRIX COMPOSITES

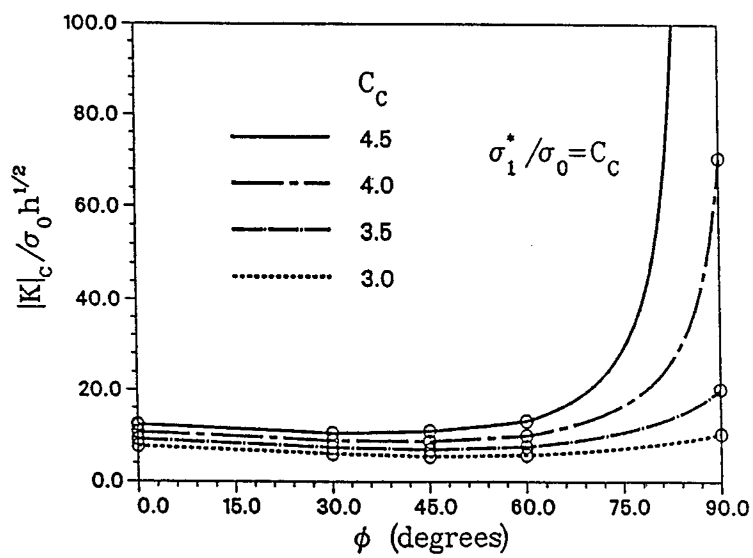


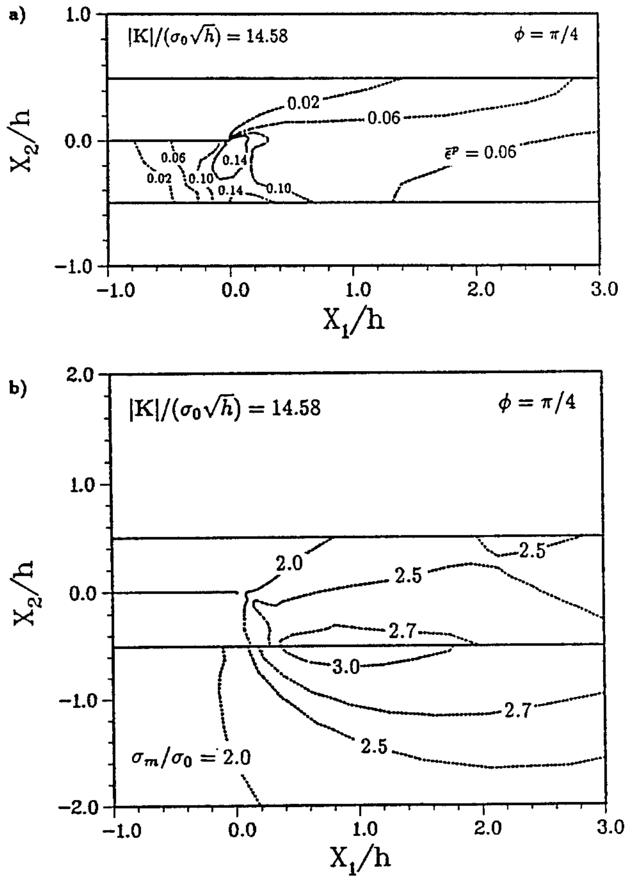


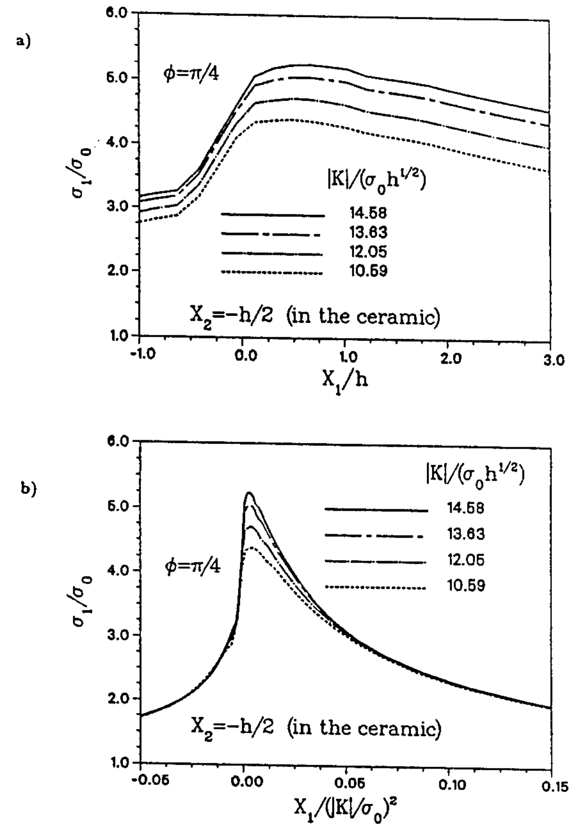
Near the crack tip failure is driven by the development of high strains and the crack may advance by void nucleation, growth and coalescence.

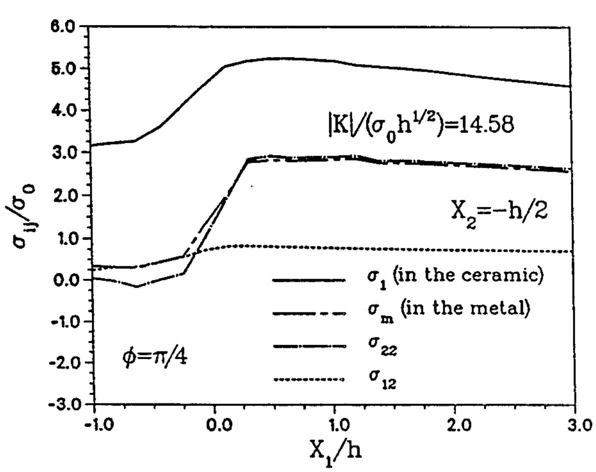
Away from the tip at distances of the order of the foil thickness failure is triggered by the elevation of the stresses. There are three potential failure mechanisms : (i) high triaxiality cavitation, (ii) interfacial deadhesion, and (iii) ceramic fracture.

The toughness of the ceramic increases with mode mixity.



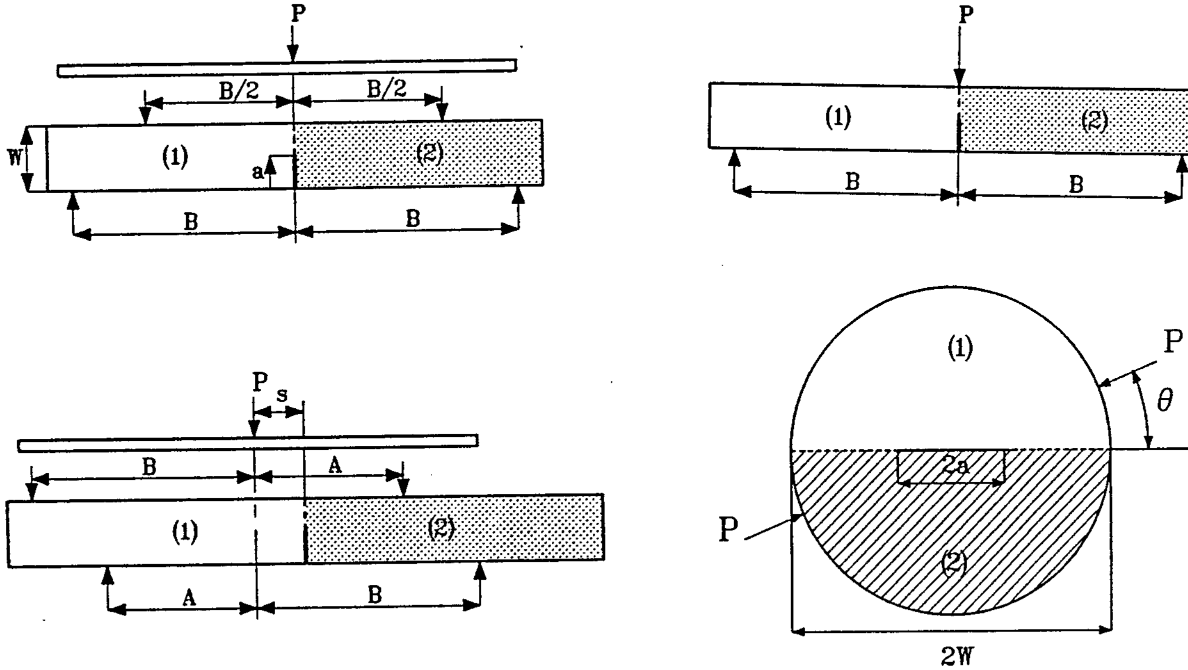




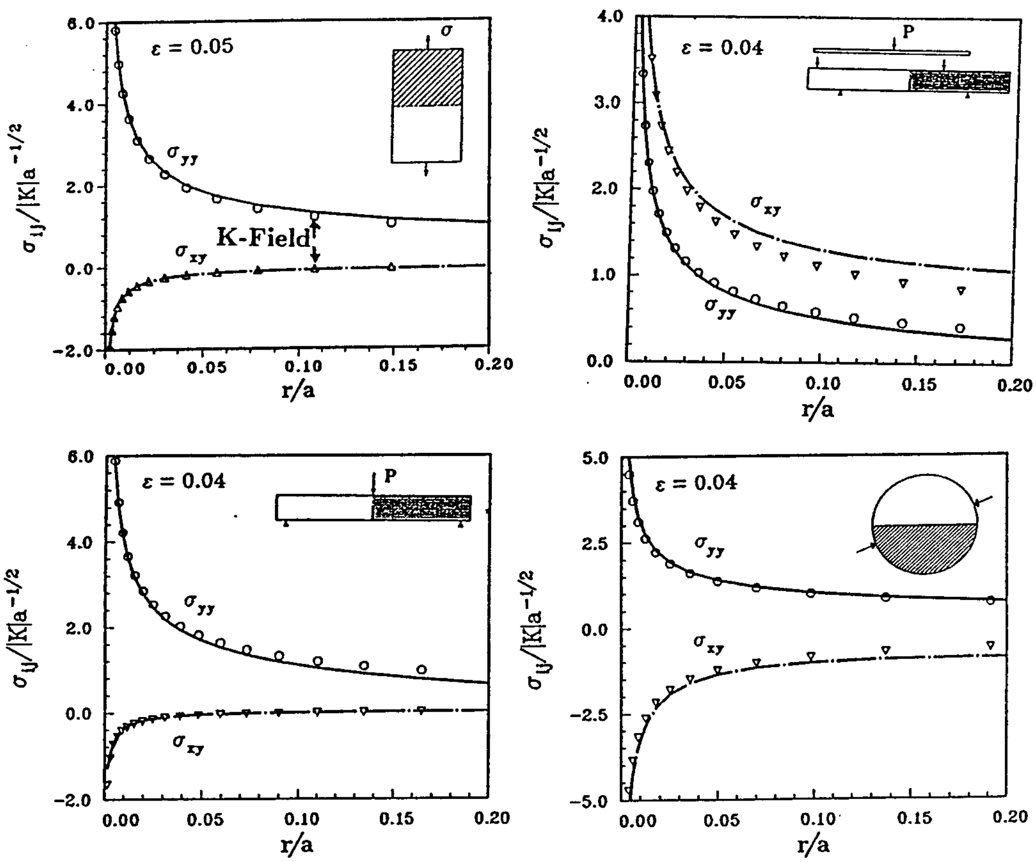


Under mixed-mode loading the plastic deformation is constrained mainly along one of the interfaces, depending on the direction of the remotely applied shear. High stresses develop along that interface and increase continuously with remote loading. The maximum tensile stress in the ceramic becomes larger than the maximum mean stress in the metal and the maximum normal interfacial traction.

TEST SPECIMENS FOR INVESTIGATING INTERFACIAL FRACTURE TOUGHNESS



COMPARISON BETWEEN FULL-FIELD SOLUTION AND K-FIELD



CALIBRATION OF SPECIMENS : Y AND ψ ARE FUNCTIONS OF GEOMETRY AND BIMATERIAL CONSTANTS, α AND β

$$K = Y T \sqrt{L} L^{-i\varepsilon} e^{i\psi}$$

FOUR POINT BEND BAR

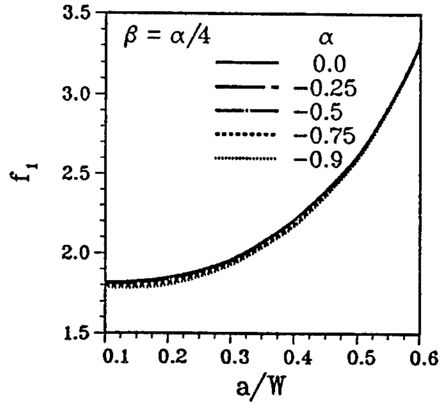
$$T = \frac{P}{W}\left[\frac{A-B}{A+B}\right]$$

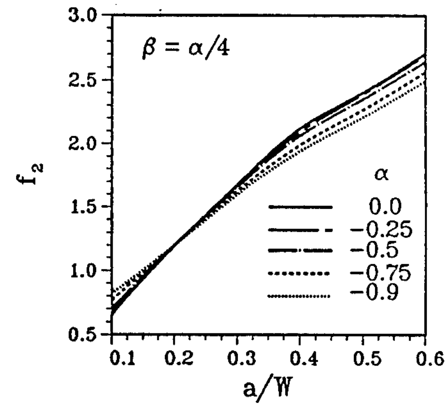
$$Y = \sqrt{Y_1^2 + Y_2^2}$$

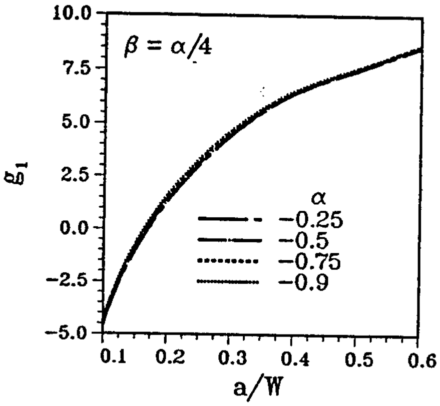
$$\psi = \tan^{-1}\left\{\frac{Y_2}{Y_1}\right\}$$

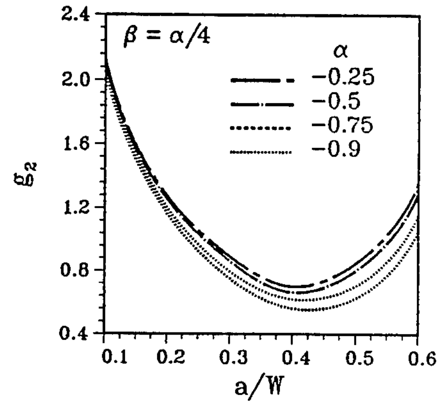
$$Y_1 = (6s/W)f_1 - 2\varepsilon g_1$$

$$Y_2 = f_2 + (12s/W)\varepsilon g_2$$



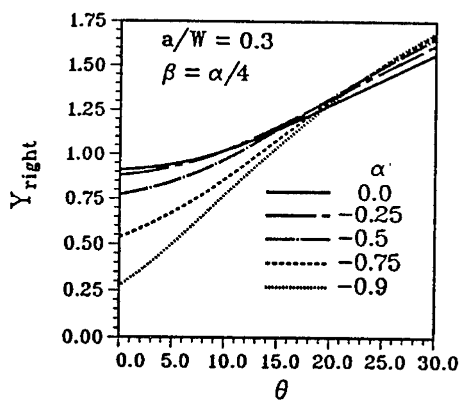


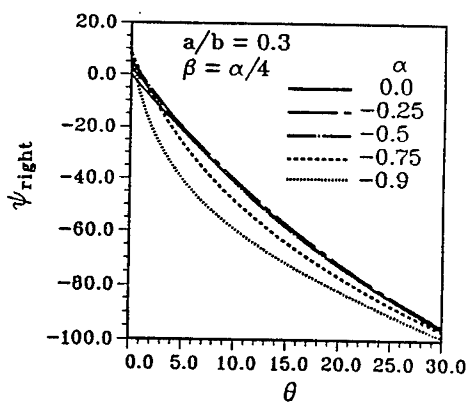


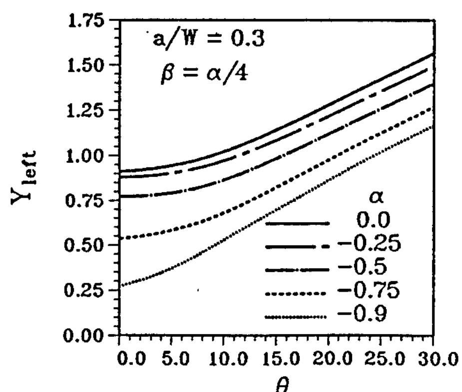


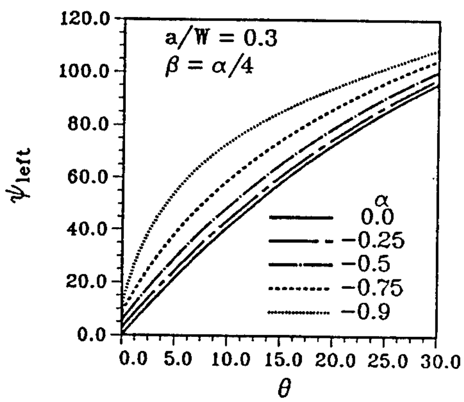
BRAZILIAN DISK

$$T = \frac{P}{2W}$$

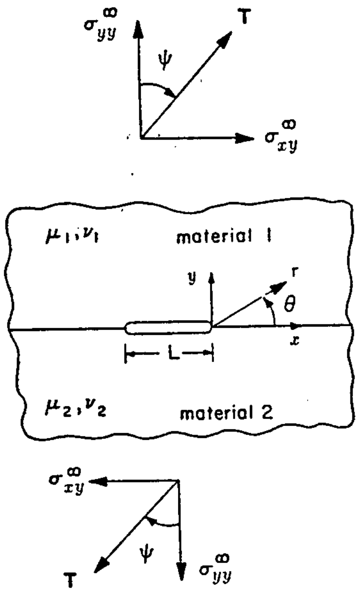


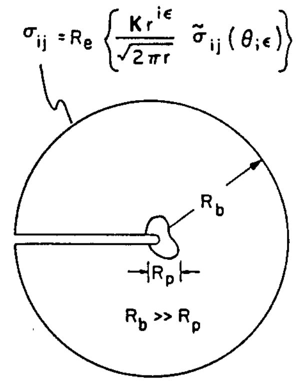


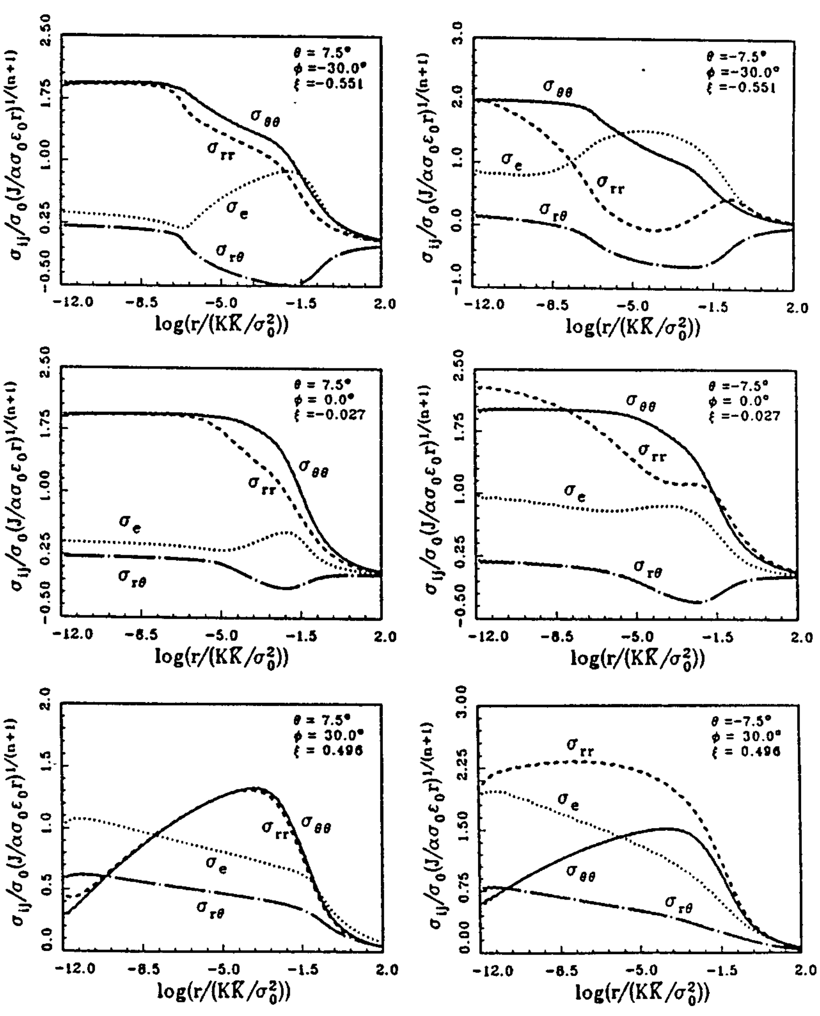




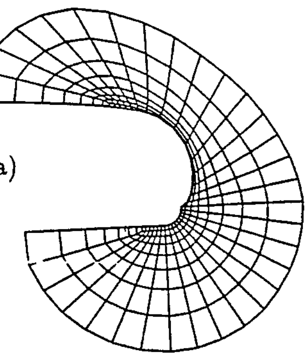
Elastic-Plastic Analysis of Cracks on Bimaterial Interfaces:



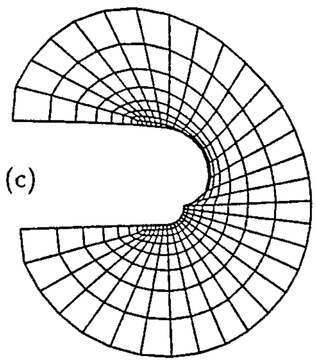




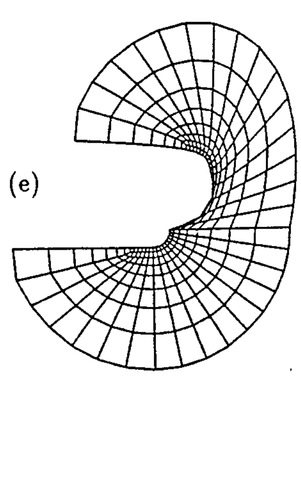
(a)



(c)



(e)



MICROMECHANICS OF RADIAL MATRIX CRACKING AND INTERPHASE FAILURE IN FIBER COMPOSITES

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ABSTRACT

Radial matrix cracking and interphase failure for transverse loading of a hexagonal array fiber composite are investigated by: 1) modeling the interphase by a layer of radial and circumferential spring elements, 2) adopting a tensile stress criterion for initiation of matrix cracking, and 3) employing a strain-energy density criterion for interphase failure. The mechanical behavior of the composite is defined in terms of geometrical, stiffness and strength parameters. Stresses on the microlevel have been calculated. Under the assumption that the failure mechanisms follow the periodicity of the composite, two scenarios related to different values of the strength parameters have been investigated. In the first scenario matrix cracking occurs first, followed by interphase failure. In the second scenario interphase failure is followed by matrix cracking. Typical results are displayed graphically.

The mechanical properties of fiber-matrix interphases significantly affect the overall stiffness and strength of a fiber-reinforced composite. For a quantitative analysis an interphase can be modeled in two ways. In a first model the interphase is considered as a thin annular layer in between the fiber and the matrix, with mechanical properties that differ from those of both the fiber and matrix materials. A number of authors have pursued this approach for a single fiber in an unbounded matrix. The thin annular layer model introduces at least three parameters, namely, the thickness of the interphase and two elastic constants. These parameters are generally very difficult to obtain. The model also gives rise to significant analytical complications when closely spaced fibers are considered and mechanical interactions between the fibers cannot be neglected.

The number of parameters can be reduced to two and the analytical complications can be decreased, by what amounts to an averaging procedure across the thickness of the annular layer. In that manner, the thickness and the elastic constants are combined into two spring constants. This simplified model can also be used if the condition between the fiber and the matrix is of the nature of contact between rough surfaces containing microcracks, voids and asperities. The spring-layer has also been used by a number of authors. The relation between the spring constants and the parameters of an annular interphase layer has been discussed elsewhere.

In this paper the compliant interphase between fibers and matrix is represented by the spring-layer model. With respect to local polar coordinates the relations between the relevant stress and displacement components may then be expressed as

$$\sigma_r^m = \sigma_r^f = k_r(u_r^m - u_r^f), \text{ if } u_r^m \geq u_r^f \quad (1)$$

$$\sigma_r^m = \sigma_r^f \text{ and } u_r^m = u_r^f, \text{ if } u_r^m \text{ not} > u_r^f \quad (2)$$

and

$$\sigma_{r\theta}^m = \sigma_{r\theta}^f = k_\theta(u_\theta^m - u_\theta^f) \quad (3)$$

where σ_r is the interfacial radial stress, $\sigma_{r\theta}$ is the interfacial shear stress, and u_r and u_θ are the displacements in the radial and the circumferential direction, respectively. Quantities with upper index 'm' and 'f' are defined in the matrix and the fiber regions, respectively. The constants k_r and k_θ are the coefficients of the springs. The addition of Eq.(2) insures that the model will not allow an unrealistic radial overlap of the two materials in the interfacial zone.

It is noted that the compliant conditions (1)-(3) include the case of perfect contact ($k_r = \infty$, $k_\theta = \infty$), when the stresses and displacements are continuous, and the case of no contact ($k_r = k_\theta = 0$) when the stresses vanish. It is also noted that for a disbond the ligament at the tip of the disbond undergoes a finite stretch when in tension, and consequently the stress remains bounded. Hence the usual problems of violently oscillating singularities that are associated with crack-tip fields for a crack in a perfectly bonded interface, do not occur for the spring-layer model.

The configurations that are considered in this paper are shown in Fig. 1 and Fig. 2. These figures show cross-sectional views of fiber-reinforced composites. The fibers, which are all of equal radius, a, are periodically spaced in a hexagonal array. It is assumed that at some large distance, the composite is subjected to uniform stresses, σ_o, applied in the x-direction. The loading direction in Fig. 1a is called a closest packing direction (CPD), and the one in Fig. 2a,

which is along the bisectrix of the CPD directions, is called the mid-closest packing direction (Mid-CPD). The spacing of the fiber centers in the Mid-CPD direction is $b\sqrt{3}$. The basic cell of the composite is a hexagonal with sides b, as shown in Figs. 1b and 2b (the regions enclosed by dashed lines). The periodicity of the composite then implies that the state of stress and deformation in the composite will be completely defined by the stresses and strains in a quarter region of a basic cell. This quarter region is shown in Fig. 1c for the CPD case and Fig. 2c for the Mid-CPD case. The boundary conditions on the external boundaries of the trapezoids have been discussed in some detail elsewhere.

Next we also include matrix cracks and interphase disbonds in the configuration, as shown in Fig. 3. It is assumed that the matrix cracks and interphase disbonds have the same periodicity as the fiber array. Hence, results can still be obtained from an analysis of the trapezoidal cell shown in Fig. 3.

An investigation of the initiation and propagation of matrix cracks and interphase disbonds must be based on appropriate criteria. Let us first consider the initiation of matrix cracks. For a perfect composite subjected to tensile stresses, numerical results show, in agreement with physical intuition, that the circumferential tensile stress at the fiber-matrix interphase is the largest tensile stress component in the matrix material. As a crack initiation criterion we therefore choose

$$\sigma_\theta \geq \sigma_\theta^{cr} \tag{4}$$

On the basis of Eq.(4), it is assumed that, in agreement with experimental observations a radial matrix crack is formed at the interphase when Eq.(4) is satisfied. It is assumed that the propagation of such a crack is governed by the fracture toughness, K_I^{cr}. Indeed, it turns out that for the far-field transverse tensile loading which is being considered here, the Mode-II stress intensity factor is negligible as compared to K_I. Hence we consider as condition for continued radial matrix cracking that

$$K_I \geq K_I^{cr} \tag{5}$$

For the generation of disbonds, as well as their propagation and arrest, it is feasible to use such criteria as critical stress, critical strain or critical strain energy density, because in the spring-layer model all these quantities are well defined near the tip of a disbond. In this paper we will employ an energy density criterion, since it combines information on the tensile and shear stresses in the interphase. The strain energy per unit interphase area is easy to calculate. We have

$$U = \frac{\sigma_r^2}{2k_r} + \frac{\sigma_{r\theta}^2}{2k_\theta} \tag{6a}$$

It is assumed that the interphase will break and form a disbond when

$$U > U^{cr} \tag{6b}$$

There are many material and geometrical parameters in the problem at hand. They may be summarized as folllows:

material parameters:

shear moduli: μ^f and μ^m

Poisson's ratios: ν^f and ν^m

interface stiffnesses: k_r and k_θ

critical stress for matrix crack initiation: σ^{cr}

fracture toughness of matrixmaterial: K_I^{cr}

critical value of interphase strain energy density: U^{cr}

loading parameters:

far-field (transverse) stress: σ_o

geometrical parameters:

fiber radius: a

fiber-center spacing: $b\sqrt{3}$

fiber volume density: $V_f = \pi a^2/\frac{3}{2}\sqrt{3}b^2$

half-length of interphase disbond: c

length of radial matrix crack: d

Numerical results for the fields of stress and deformation in the trapezoidal cells shown in Fig. 1c and 2c have been obtained by the use of the boundary element method. The details of the numerical approach can be found elsewhere. The numerical calculations were carried out for solids with the following material properties:

matrix: $\mu^m = 14.2$ Msi $\quad \nu^m = 0.22$

fibers: $\mu^f = 30$ Msi $\quad \nu^f = 0.22$

The results will actually apply for any pair of solids, which have the stated Poisson's ratios, and whose ratio of the shear moduli is the same as for the above materials. The interphase constants k_r and k_θ were rendered dimensionless by division by μ^m/a, where a is the radius of the fibers.

$$k_1 = k_r/(\mu^m/a), \quad k_2 = k_\theta/(\mu^m/a) \tag{7a,b}$$

In the computations, the two interphase constants were taken equal in magnitude: $k = k_1 = k_2$. The following values were considered $k = .2$, $k = 1$, $k = 10$ and $k = \infty$. These values represent an increasingly stiff interphase, with $k = \infty$ corresponding to a perfect bond. The area of the trapezoid in either Fig. 1c or Fig. 2c is $3\sqrt{3}b^2/8$. The fiber volume ratio, V_f, then becomes

$$V_f = \frac{\pi a^2}{4} \Big/ \frac{3\sqrt{3}b^2}{8} \tag{8}$$

Calculations were carried out for a perfect

composite as well as for composites containing radial matrix cracks and/or interphase disbonds. The configurations are shown in Figs. 1-3. It should be noted that the cracked configuration applies to a composite for which all basic cells contain the same symmetrically oriented cracks and disbonds. The lengths of the boundary elements for all calculations were chosen as 0.04a or smaller.

Appropriate modeling of the fiber-matrix interphase together with criteria for matrix cracking and interphase failure have yielded analytical and numerical results for the mechanical behavior and failure sequences of a fiber-composite. Three cases have been considered: 1) no failure, 2) radial matrix cracking first, and 3) interphase failure first. For specified values of the geometrical, stiffness and strength parameters it can be determined which case will apply. Figures 4-12 display the results.

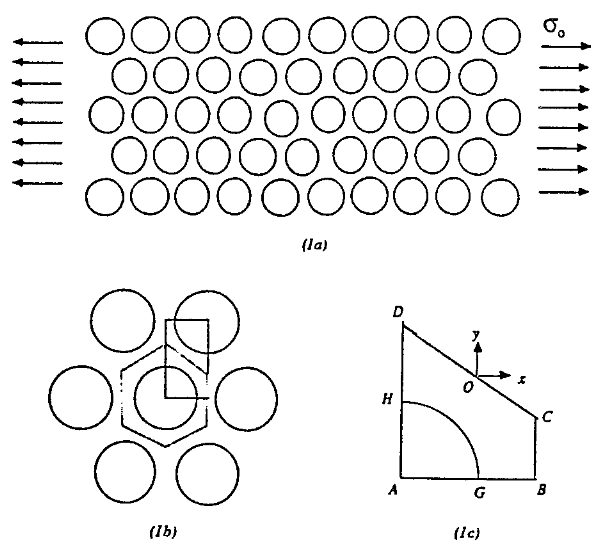

Fig. 1. (a) Hexagonal array subjected to far-field uniform tensile stress in the closest packing direction (CPD); (b) Basic cell; (c) Trapezoidal domain for numerical calculations

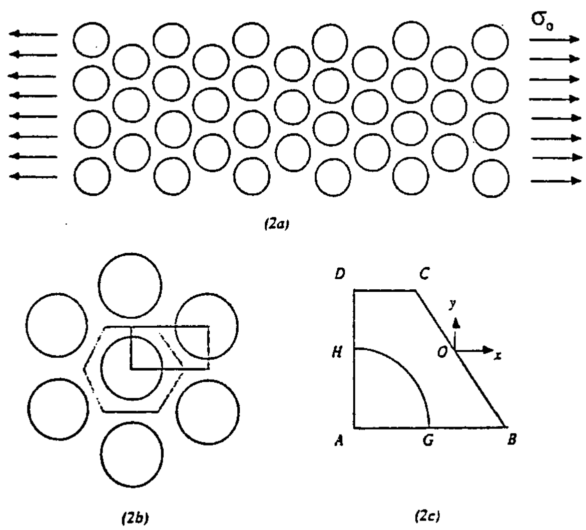

Fig. 2. (a) Hexagonal array subjected to far-field uniform tensile stress in the mid-closest packing direction (Mid-CPD): (b) Basic cell; (c) Trapezoidal domain for numerical calculations

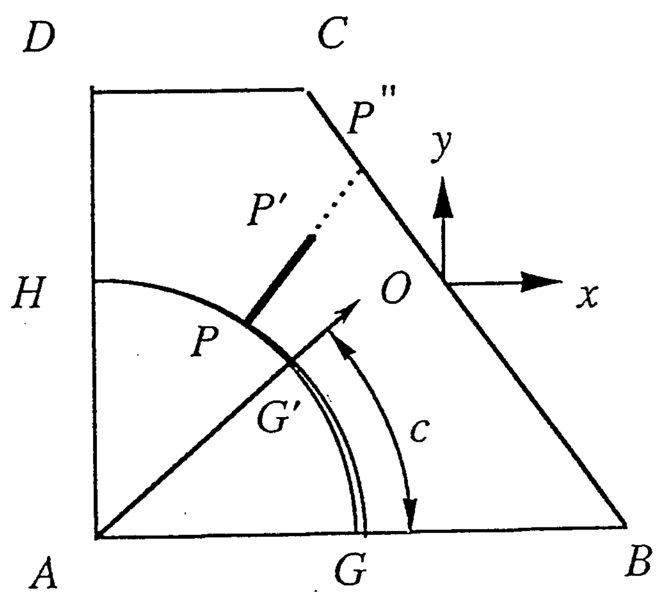

Fig. 3. Trapezoidal domain with radial matrix crack and interphase disbond

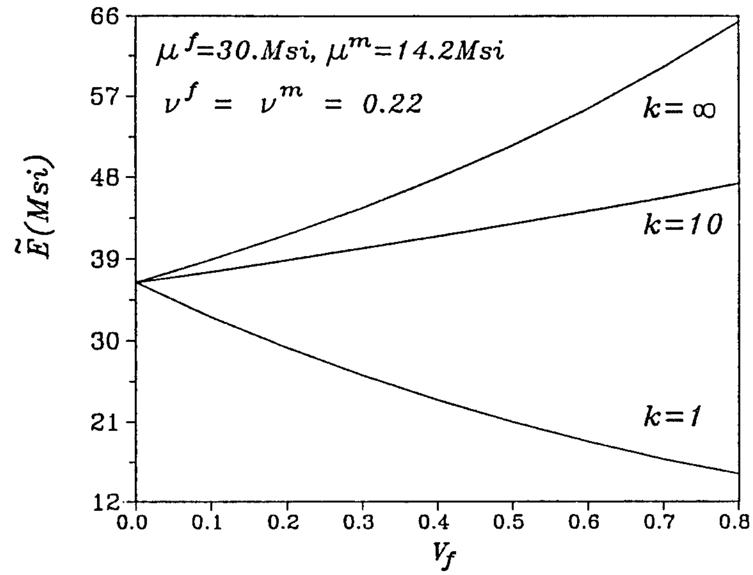

Fig. 4. Transverse elastic constant versus fiber volume fraction for various values of the interphase constant

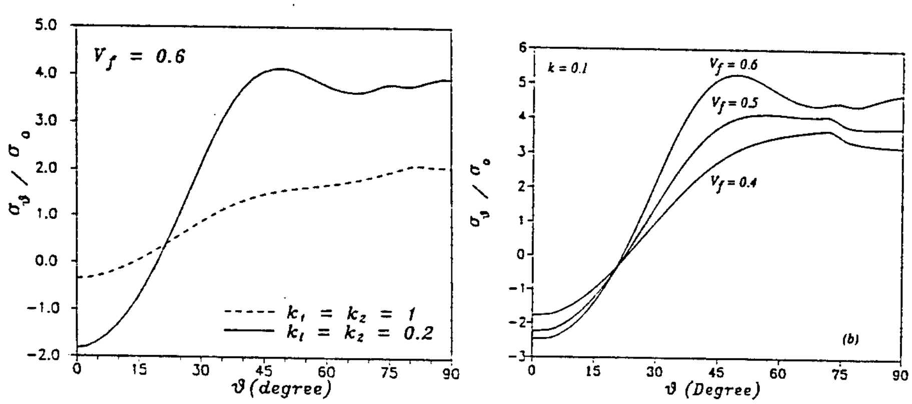

Fig. 5. Circumferential stress at the matrix side of the interphase

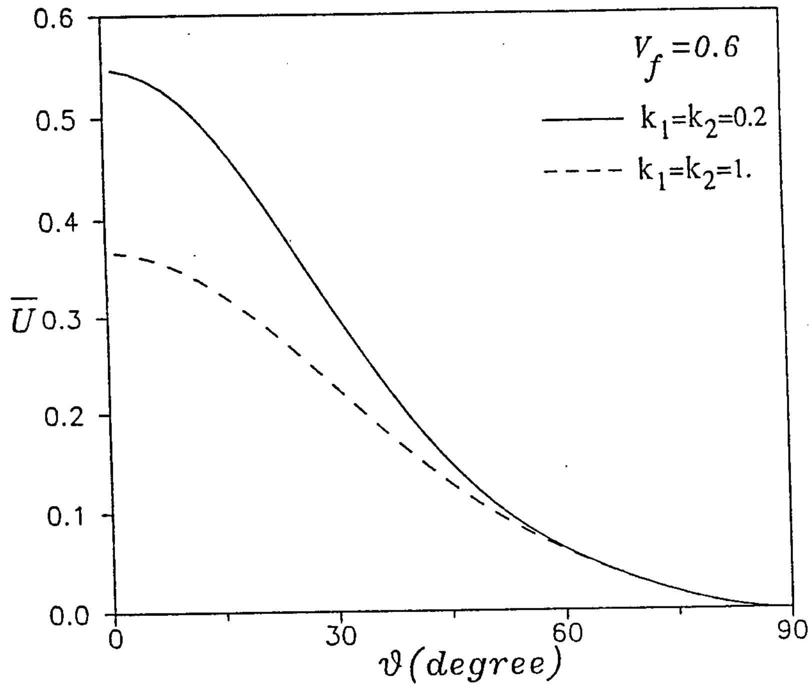

Fig. 6. Interphase strain energy density

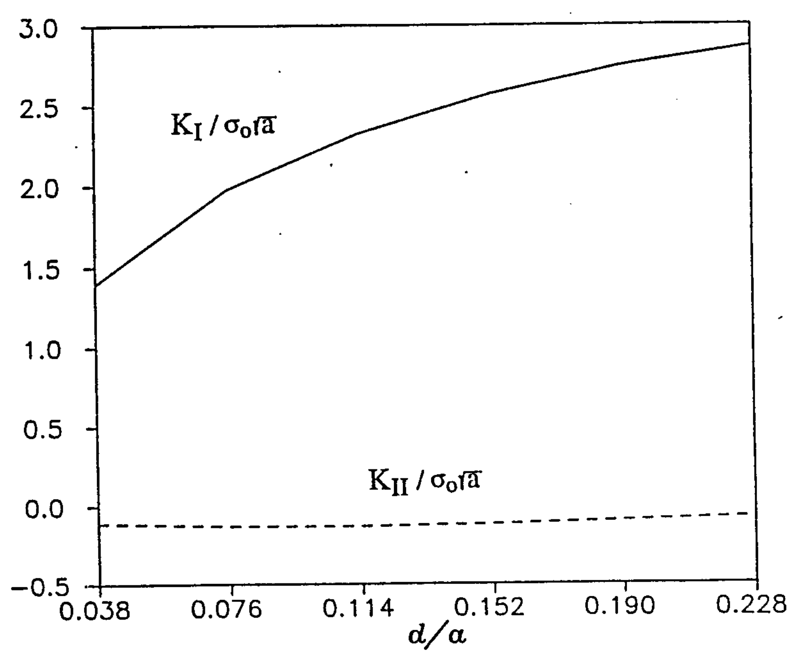

Fig. 7. Stress intensity factors versus the length of the radial matrix crack

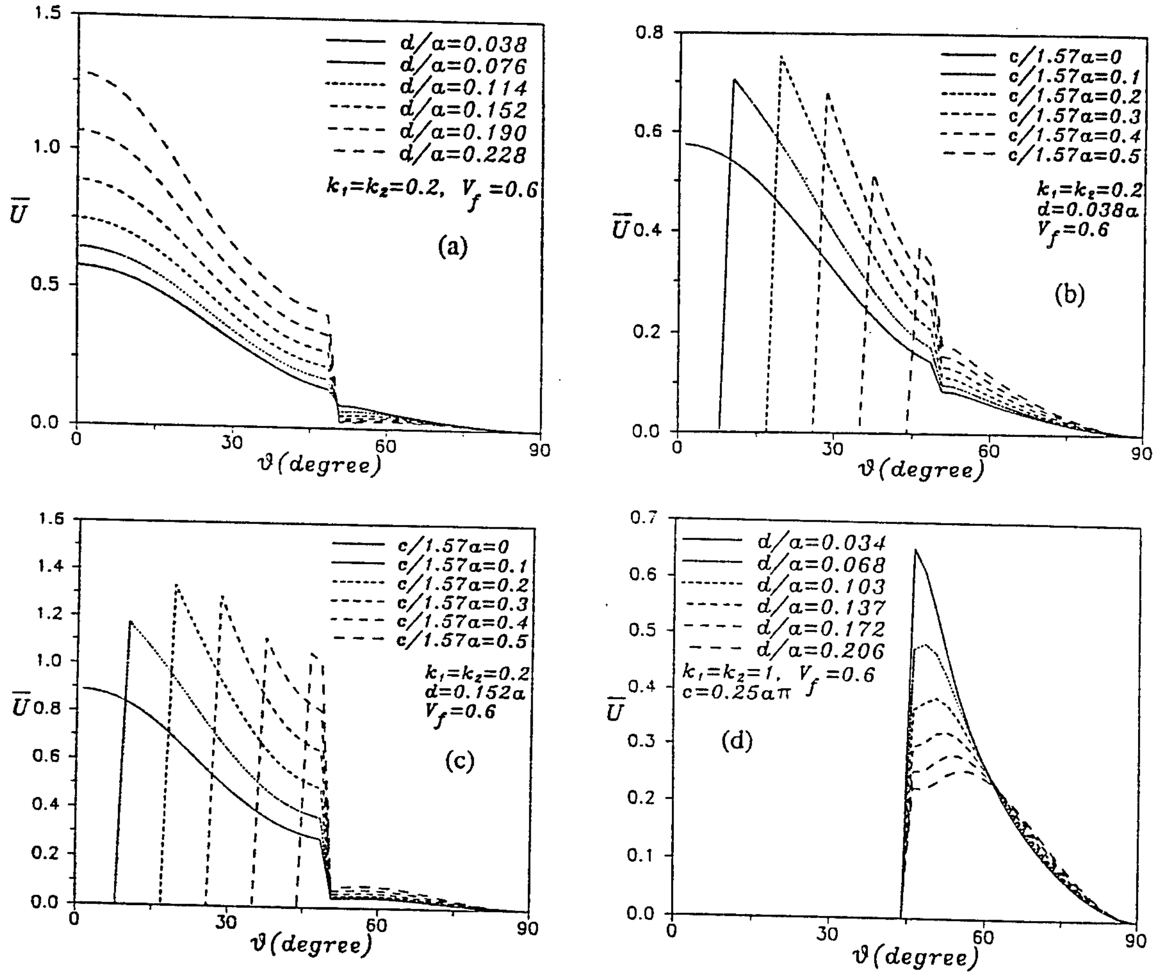

Fig. 8. Interphase strain energy density for various lengths of the radial matrix crack and/or the interphase disbond

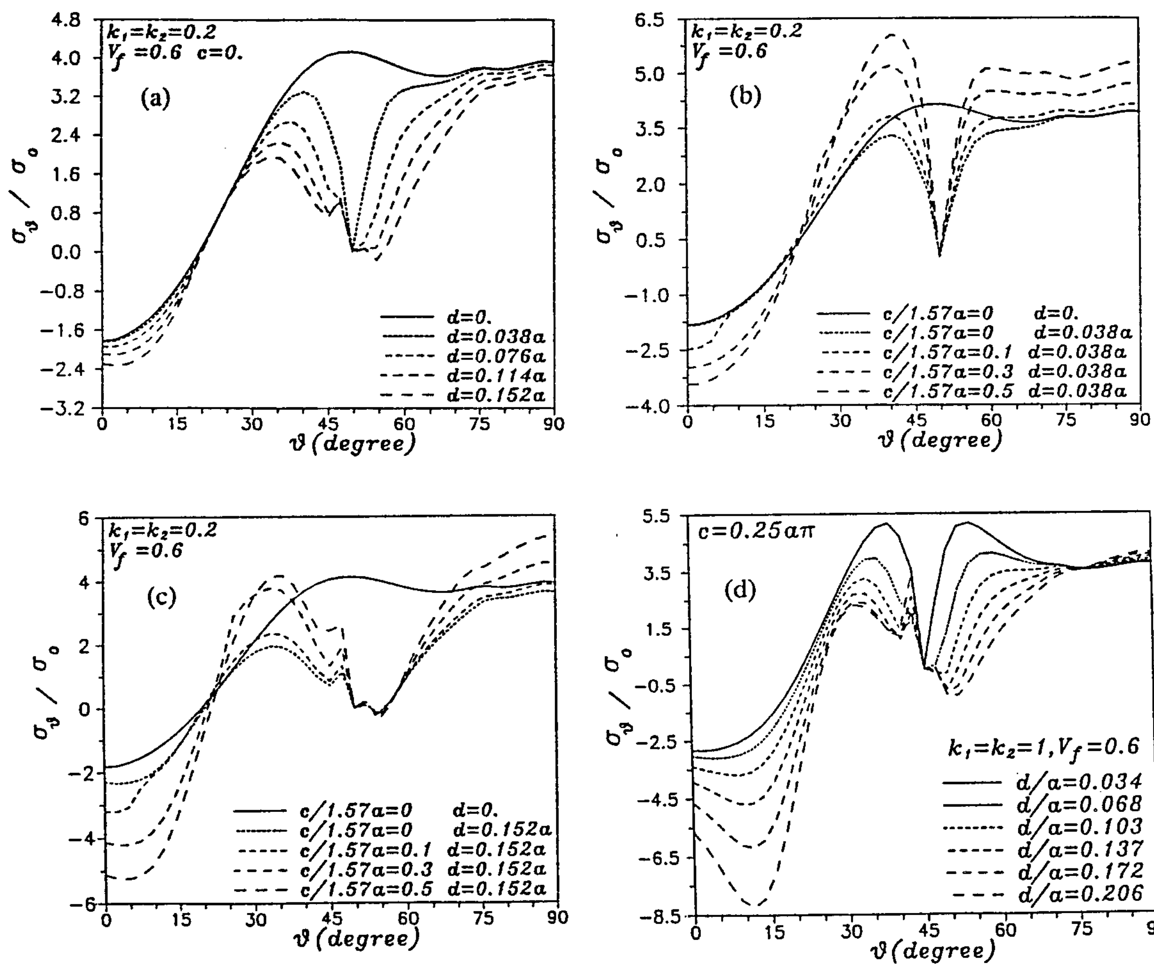

Fig. 9. Circumferential stress at the matrix side of the interphase for various lengths of the radial matrix crack and/or the interphase disbond

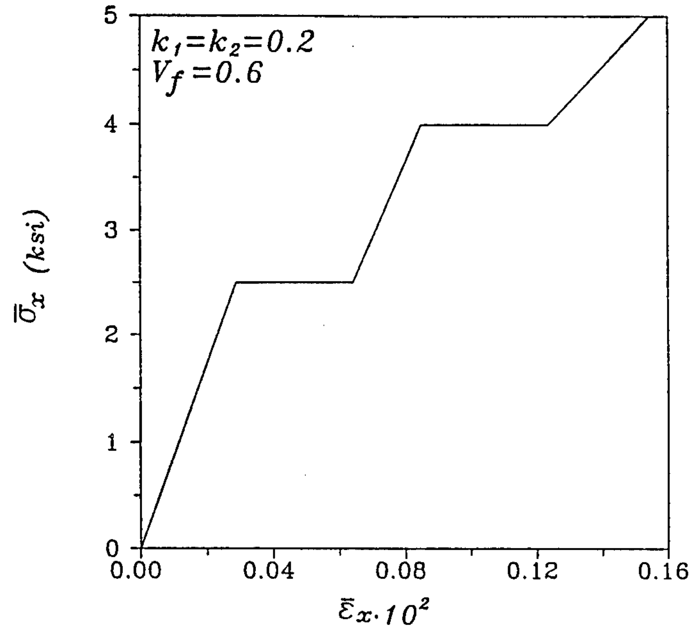

Fig. 10. Effective stress-strain relation when radial matrix cracking is followed by interphase disbonding

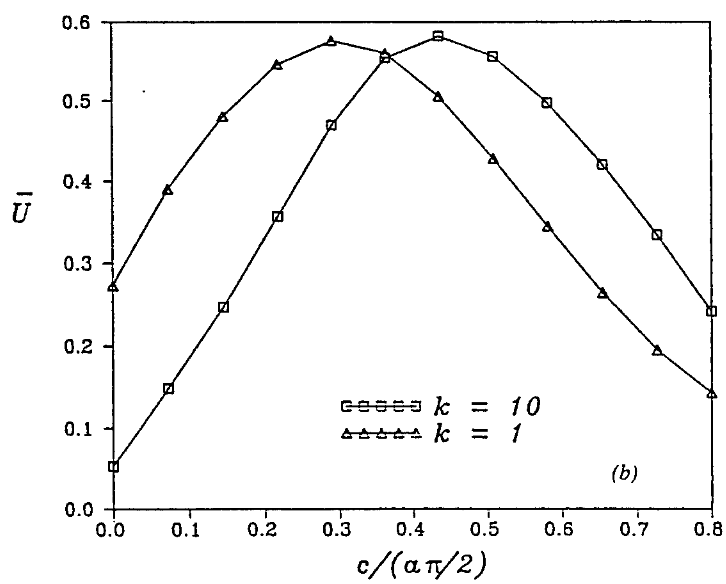

Fig. 11. Interphase strain energy density at the tip of the disbond versus the length of the disbond

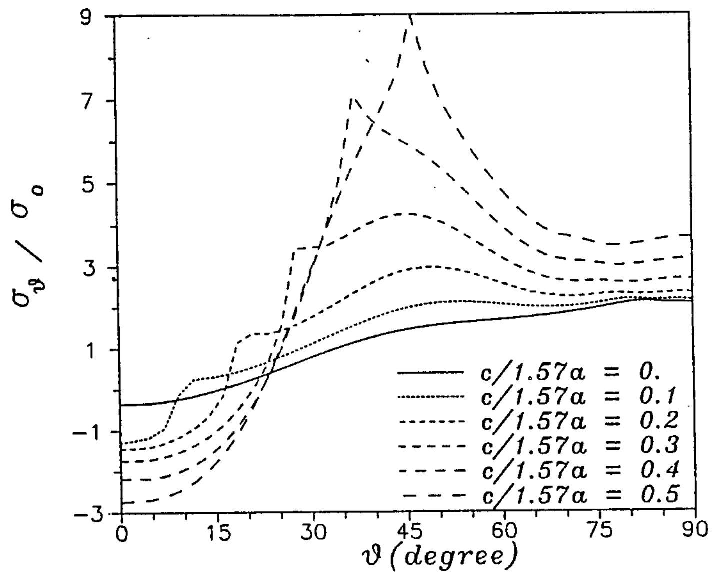

Fig. 12. Circumferential stress at the matrix side of the interphase for various lengths of the interphase disbond

Mechanics of Thick Composites

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Abstract:

In the framework of 3 dimensional elasticity, for a thick laminate, the strains $\epsilon_{\alpha\beta}$ (α = 1,2 in plane) and stresses σ_{3i} (i = 1,2,3 ; i = 3 the thickness coordinate) are continuous; while ϵ_{3i} and $\sigma_{\alpha\beta}$ are discontinuous, at lamina inter faces. The validity and limitations of the tso-called classical laminate plate theory, the higher-order-shear-deformation theories, and the layer-wise linear multi-director theories are critically examined. The need for simple, reliable, and accuracte nonlinear dynamical and vibration analyses of thick lamintes is discussed, and the progress made so far is reviewed.

The high $E_{\alpha\beta}/G_{\alpha 3}$ ratios of the laminate cause serious warping of the composite section and influence the thickness stresses significantly. At the free edges, the high interlaminar stresses cause initiation and propagation of delamination and damage processes. The influence of the stacking sequence on the interlaminar stresses, as well as on the magnitudes of the frequencies of free vibration is discussed; and a need for optimization if the stacking sequence with respect to these criteria is pointed out.

The prediction of the behavior of a thick section composite *under compression* is as yet an unmet challenge. The failure modes of fiber microbuckling, kink band formation, delamination, and the influences of the environment are yet to be fully understood.

Also, the detailed studies of thick section composites subject to medium velocity impact, taking into account all significant physical processes of damage, are lacking. The sensitivity of the behavior of thick composites to variations in manufacturing quality is also a subject of research concern.

The possibility of designing *smart structures*, by embedding sensor/actuator surfaces/fibers (shape memory alloys) in a composite laminate, is discussed. These embedded sensors can be used in vibration control as well as in the control of failure processes such as delamination. Also, the feasibility of *acoustic tailoring* of a composite structure through *designing* its vibration characteristics, is discussed in detail. The need for developing *smart-algorithms* for analyzing the fluid-structure interaction problems is discussed.

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RELATIVE SHOCK PERFORMANCE OF THICK COMPOSITE NAVAL STRUCTURES

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ABSTRACT

The use of composite materials in naval structures requires that the material/structural design meet many severe criteria. One of the most severe of these is dynamic loading due to waterborne shock waves. An on-going program to investigate the dynamic response of several composite material systems is described. Small-scale cylindrical models fabricated from four different material systems are tested underwater using small explosive charges to induce rapidly applied pressure loads[1]. The material systems investigated include S2 glass fibers in epoxy and thermoplastic matrices, AS4 fibers in a thermoplastic matrix and Celion 6K carbon fibers in an epoxy matrix. The models are instrumented and strain response is recorded during the tests. Extensive post-test C-scan inspection is performed to detect the onset of damage during the test series. The strain response and damage accumulation of the various models are compared to obtain first-order determinations of the relative capabilities of these systems.

Finite element analyses of the dynamic tests are performed using the ADINA-S [2,3] finite element code which has the capability to determine the spatially varying applied pressure due to detonation of an explosive underwater as well as the interaction pressure generated between the structure and surrounding medium. These analyses are providing generally good agreement between measured and calculated global strain response of the structure. In the more severe tests in which material nonlinearity and possibly damage become more pronounced, the agreement between linear elastic orthotropic analysis and experiment is less satisfactory.

As an extension of this program, the effect of model scale is being investigated by testing an S2 glass/epoxy cylinder of exactly twice the dimensions of the smaller models. By varying all the test parameters using Hopkinson scaling laws, the same strain magnitudes obtained in the small model tests are expected for the same severity test in the larger model.

Finally, an explosively loaded flat plate composite element test is described. This test and analysis program is an effort to investigate the response of thick section (1 in. and greater) composite elements which are attached to a fixture and dynamically loaded. The loading is generated by detonating a sheet explosive in a water column situated over the element and fixture. Calibration testing using aluminum plates has shown the response of the element to be axisymmetric which will simplify the supporting analyses. It is planned that this composite element test will provide a means to test thick composite material systems under extremes of strain magnitude and rate and serve as a means to validate analytic procedures including the application of appropriate failure criteria.

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RELATIVE SHOCK PERFORMANCE OF THICK COMPOSITE NAVAL STRUCTURES

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OBJECTIVES

- DEVELOP EXPERIMENTAL TECHNIQUES REQUIRED TO ASSESS THE SHOCK RESPONSE AND DAMAGE RESISTANCE OF THICK COMPOSITE MATERIALS AND STRUCTURAL CONCEPTS
- DEVELOP THE ANALYTICAL CAPABILTIES REQUIRED TO REPRODUCE AND PREDICT THE RESPONSE OF THESE COMPOSITE STRUCTURAL CONCEPTS UNDER SHOCK LOADING
- BEGIN THE DEVELOPMENT OF RATIONAL MATERIAL AND STRUCTURAL DYNAMIC RESPONSE CRITERIA FOR COMPOSITES IN NAVAL STRUCTURAL APPLICATIONS

TECHNICAL APPROACH

- REVIEW LITERATURE/ON-GOING INVESTIGATIONS
- PARTICIPATE IN DYNAMIC MATERIAL TEST DEVELOPMENT
- DEVELOP STRUCTURAL MODEL TEST USING 8-IN. CYLINDER
- PERFORM AND REFINE FINITE ELEMENT ANALYSES
- EXTEND MODEL TESTS AND ANALYSES TO OTHER MATERIALS
- INVESTIGATE EFFECT OF SCALE ON DYNAMICALLY LOADED THICK COMPOSITE STRUCTURES
- DEVELOP FLAT PLATE TYPE STRUCTURAL ELEMENT TEST FOR THICK COMPOSITE PANELS

RESULTS OF LITERATURE REVIEW

FIBER DIRECTION

- GLASS FIBER COMPOSITES ARE STRAIN RATE SENSITIVE

  TENSILE STRENGTH 24 TO 200% INCREASE
  COMPRESSIVE STRENGTH 55 TO 100% INCREASE
  TENSILE MODULUS 40 TO 150% INCREASE

- CARBON FIBER COMPOSITES ARE MINIMALLY RATE SENSITIVE

  30% INCREASES AND DECREASES MEASURED IN MATERIAL PROPERTIES

TRANSVERSE DIRECTION AND RESINS

- STRAIN RATE EFFECTS ARE UNCLEAR

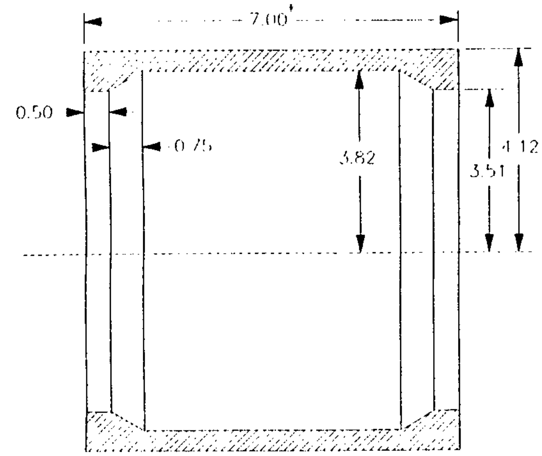
  DECREASES OF 30% TO INCREASES OF 200% MEASURED

8 IN. DIAMETER CYLINDER MODELS

- ALL MODELS WERE ORIGINALLY FABRICATED WITH 7 IN. INSIDE DIA. AND 0.6 TO 0.63 IN. THICKNESS AND THEN MACHINED TO TEST CONFIGURATION
- STRUCTURAL MODIFICATIONS WERE OPTIMIZED USING STATIC FINITE ELEMENT ANALYSES
- TEST CONFIGURATION WAS DESIGNED TO INDUCE DAMAGE IN TEST SECTION AND NOT AT END CLOSURES
- MATERIALS INCLUDE CELION 6K/E707, S2/ERL 2258, S2/JII AND AS4/JII
- $(90_2 / 0)_N$ LAMINA ORIENTATION USED IN ALL CYLINDERS

STRUCTURAL CONFIGURATION

† dimensions in inches



CYLINDER DYNAMIC TEST METHOD

- MULTIPLE TESTS ON EACH MODEL AT INCREASING PRESSURE LOADINGS
- TEN CHANNELS OF STRAIN DATA RECORDED DURING EACH TEST
- SHALLOW SUBMERGENCE (SHOCK WAVE LOADING ONLY)
- VISUAL AND C-SCAN INSPECTION AFTER EACH TEST
- FAILURE ANALYSIS USING SEM PLANNED AFTER DAMAGE NOTED BY NDE

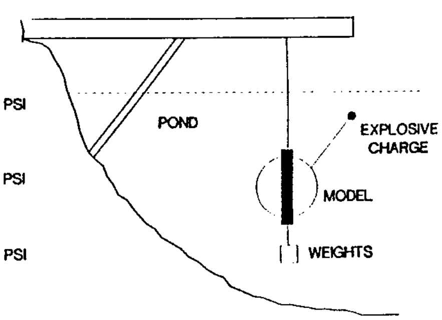
M SERIES TEST GEOMETRIES

CHARGE WEIGHT = 4 GRAMS PENTOLITE
DEPTH OF CHARGE = 4-5 INCHES
MAXIMUM BUBBLE RADIUS = 10 INCHES

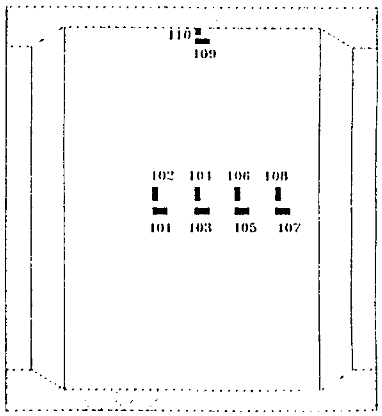
SHOT 1
STANDOFF = 2.4 FEET
PEAK PRESSURE = 1330 PSI

SHOT 2
STANDOFF = 0.67 FEET
PEAK PRESSURE = 6110 PSI

SHOT 3
STANDOFF = 0.51 FEET
PEAK PRESSURE = 8340 PSI

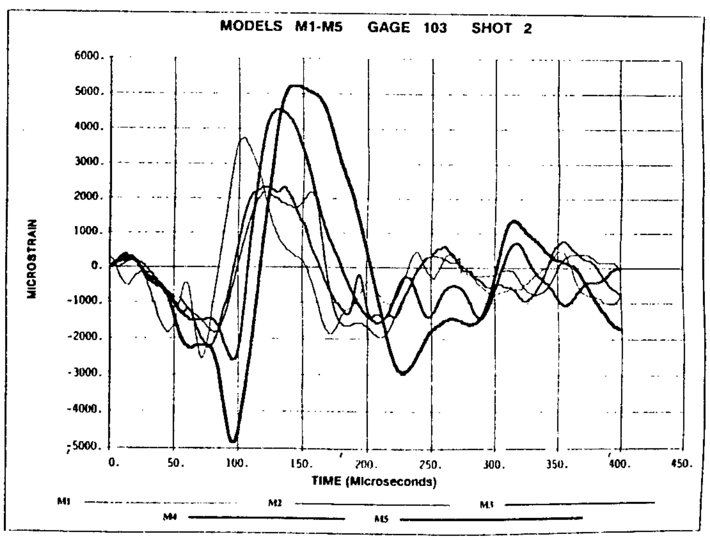


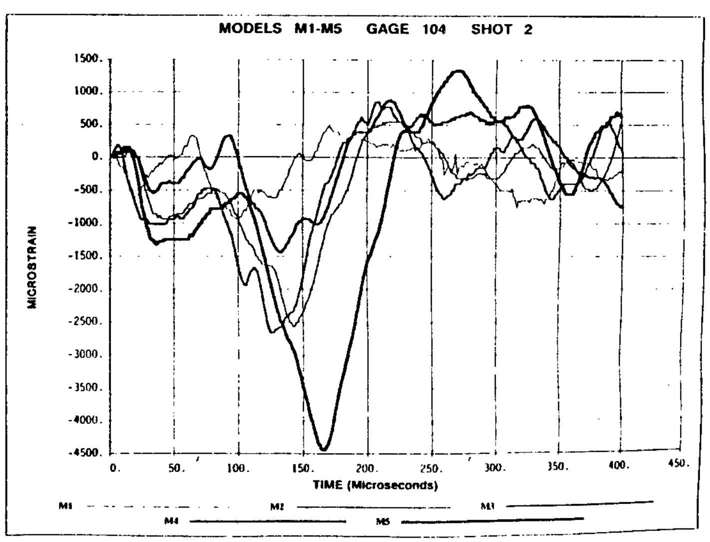
M-SERIES INSTRUMENTATION LOCATION

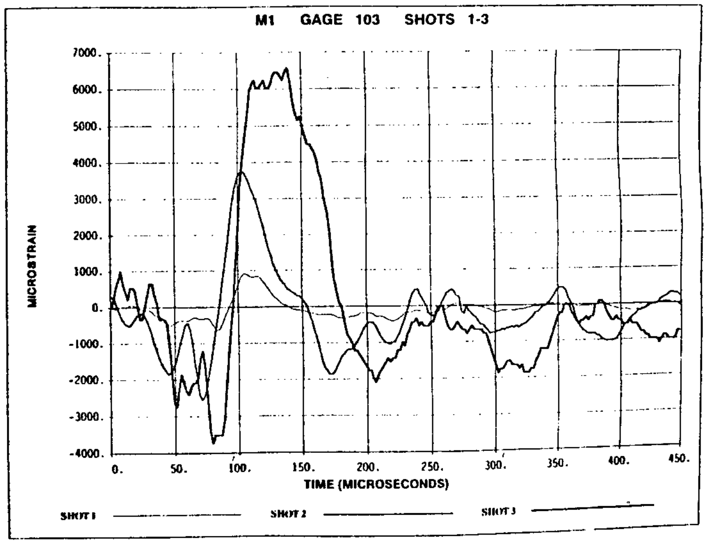


DYNAMIC TEST RESULTS

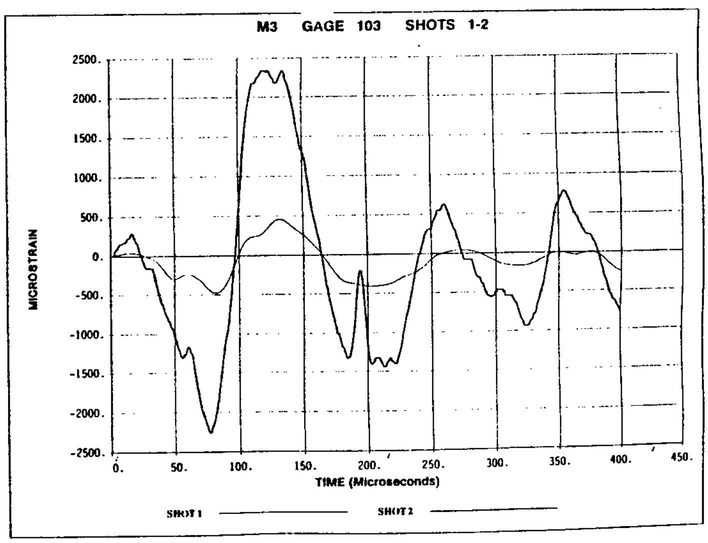
MODEL	MATERIAL	PEAK STRAINS		DAMAGE
1	CELION/E707	920	-750	None
		3700	-2600	None
		6400	-3000	Delam.
2	S2/ERL 2258	420	-940	None
		2200	-2500	None
3	S2/ERL 2258	460	-930	None
		2300	-2600	None
4	AS4/J2	750	-890	None
		4600	-2600	None
		5300	-3400	None
5	S2/J2	1250	-1580	None
		5200	-4800	None
		6800	-6300	None



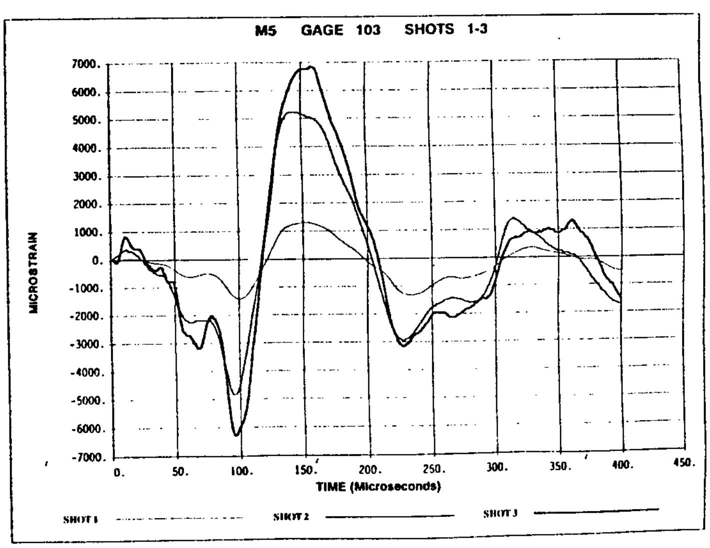




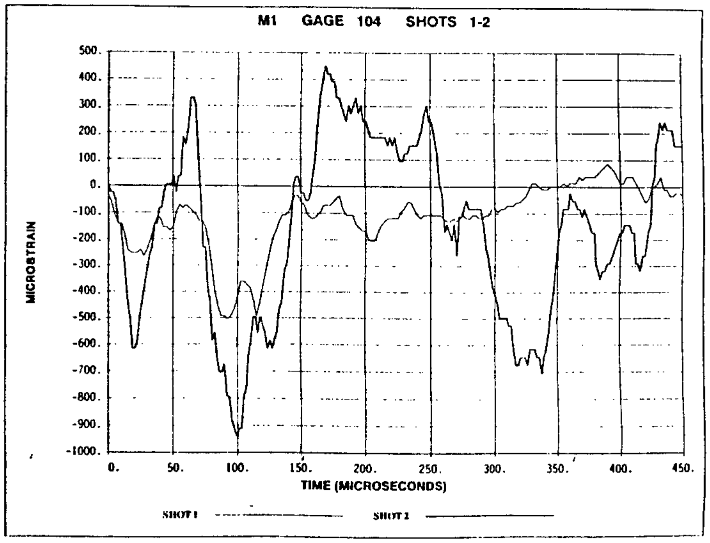
M1 GAGE 103 SHOTS 1-3
MICROSTRAIN
7000.
6000.
5000.
4000.
3000.
2000.
1000.
0.
-1000.
-2000.
-3000.
-4000.
0.
50.
100.
150.
200.
250.
300.
350.
400.
450.
TIME (MICROSECONDS)
SHOT 1
SHOT 2
SHOT 3



M3 GAGE 103 SHOTS 1-2
MICROSTRAIN
2500.
2000.
1500.
1000.
500.
0.
-500.
-1000.
-1500.
-2000.
-2500.
0.
50.
100.
150.
200.
250.
300.
350.
400.
450.
TIME (Microseconds)
SHOT 1
SHOT 2



M5 GAGE 103 SHOTS 1-3
MICROSTRAIN
7000.
6000.
5000.
4000.
3000.
2000.
1000.
0.
-1000.
-2000.
-3000.
-4000.
-5000.
-6000.
-7000.
0.
50.
100.
150.
200.
250.
300.
350.
400.
450.
TIME (Microseconds)
SHOT 1
SHOT 2
SHOT 3

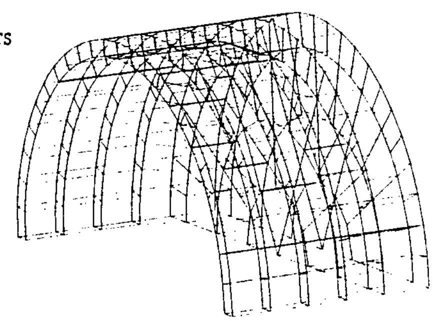


FINITE ELEMENT ANALYSIS
ADINA-S FLUID/STRUCTURE INTERACTION CODE

- 3-D CONTINUUM ELEMENTS USED IN ALL ANALYSES
- LINEAR ELASTIC ORTHOTROPIC MATERIAL PROPERTIES (STATIC)
- LAGRANGIAN FORMULATION AND IMPLICIT TIME INTEGRATION
- SPHERICAL INCIDENT WAVE (EXPONENTIAL TIME AND INVERSE STANDOFF DECAY) CALCULATED BY ADINA-S FOR GIVEN TEST GEOMETRY
- INTERACTION PRESSURE CALCULATED AT EACH TIME STEP USING DAA AND ADDED TO INCIDENT PRESSURE TO GIVE DYNAMIC PRESSURE LOADING

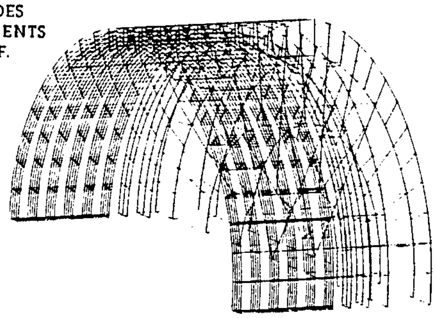
FINITE ELEMENT IDEALIZATION

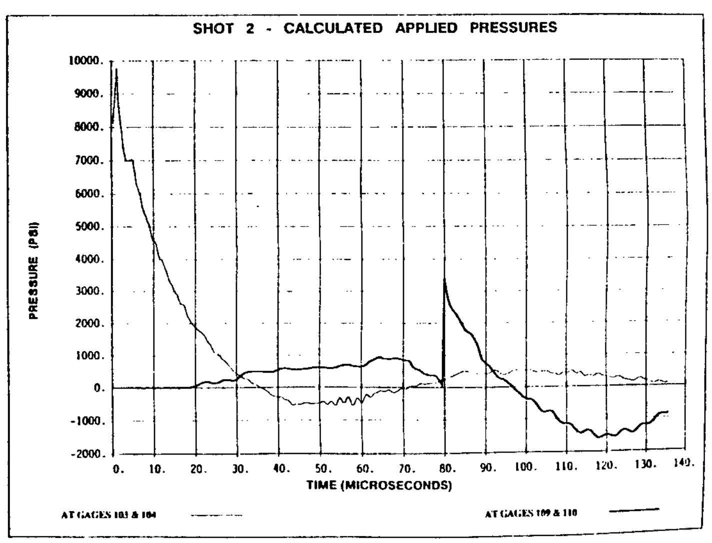
1130 NODES
141 ELEMENTS
1942 D.O.F.

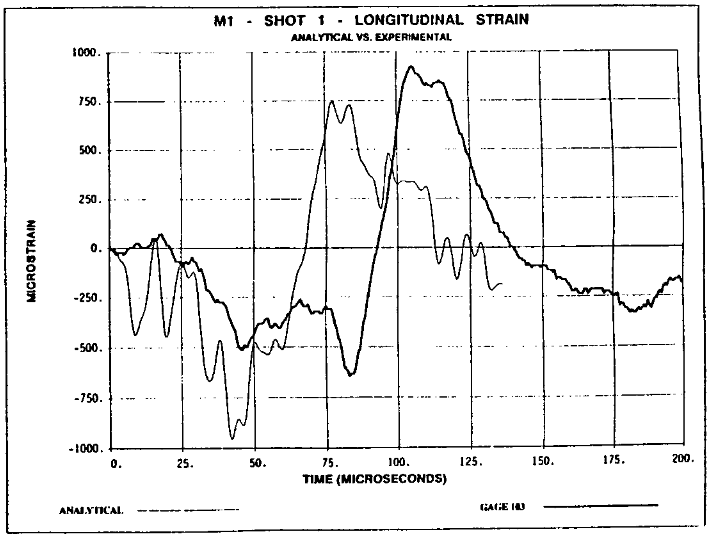


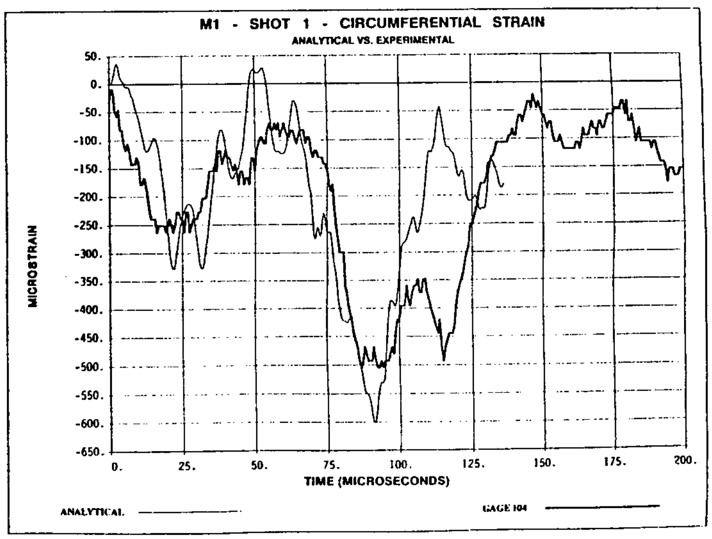
REFINED FINITE ELEMENT IDEALIZATION

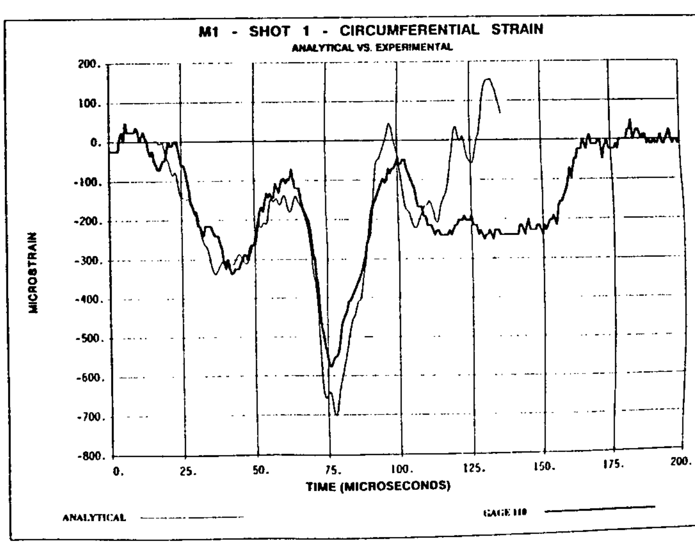
2733 NODES
645 ELEMENTS
7576 D.O.F.

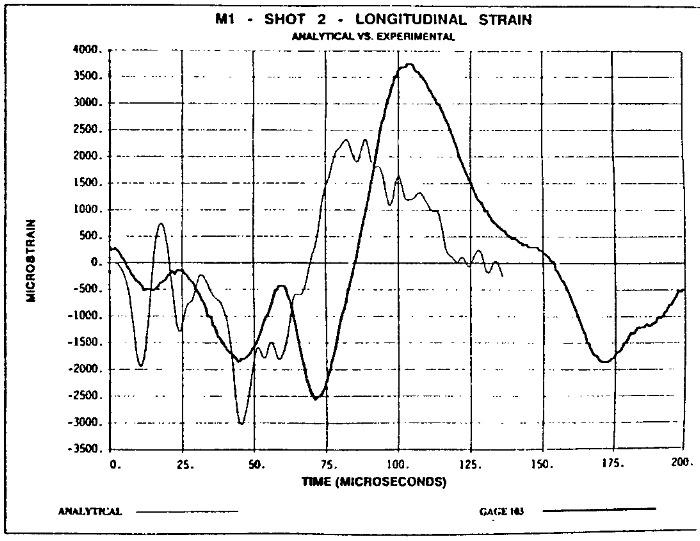


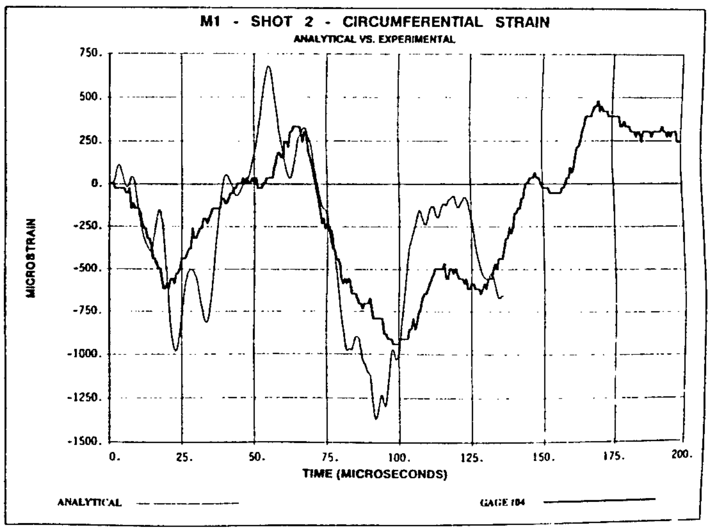


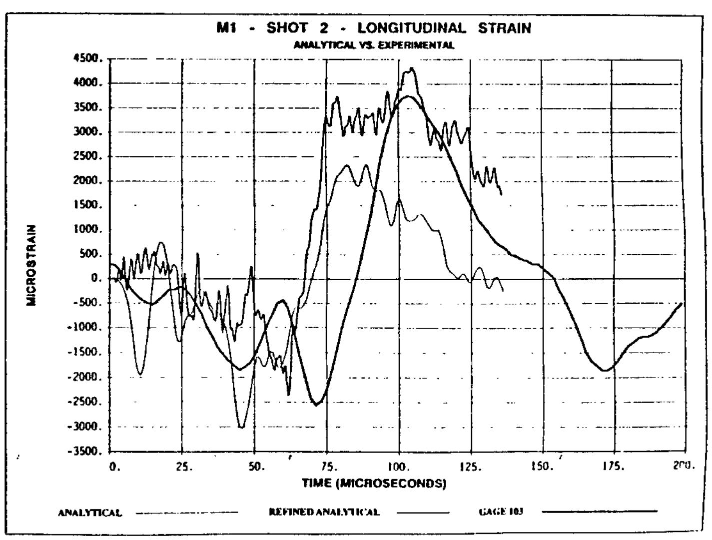

M1 - SHOT 1 - LONGITUDINAL STRAIN
ANALYTICAL VS. EXPERIMENTAL
MICROSTRAIN
1000.
750.
500.
250.
0.
-250.
-500.
-750.
-1000.
0.
25.
50.
75.
100.
125.
150.
175.
200.
TIME (MICROSECONDS)
ANALYTICAL
GAGE 103


M1 - SHOT 1 - CIRCUMFERENTIAL STRAIN
ANALYTICAL VS. EXPERIMENTAL
MICROSTRAIN
50.
0.
-50.
-100.
-150.
-200.
-250.
-300.
-350.
-400.
-450.
-500.
-550.
-600.
-650.
0.
25.
50.
75.
100.
125.
150.
175.
200.
TIME (MICROSECONDS)
ANALYTICAL
GAGE 104


M1 - SHOT 1 - CIRCUMFERENTIAL STRAIN
ANALYTICAL VS. EXPERIMENTAL
MICROSTRAIN
200.
100.
0.
-100.
-200.
-300.
-400.
-500.
-600.
-700.
-800.
0.
25.
50.
75.
100.
125.
150.
175.
200.
TIME (MICROSECONDS)
ANALYTICAL
GAGE 110


M1 - SHOT 2 - LONGITUDINAL STRAIN
ANALYTICAL VS. EXPERIMENTAL
MICROSTRAIN
4000.
3500.
3000.
2500.
2000.
1500.
1000.
500.
0.
-500.
-1000.
-1500.
-2000.
-2500.
-3000.
-3500.
0.
25.
50.
75.
100.
125.
150.
175.
200.
TIME (MICROSECONDS)
ANALYTICAL
GAGE 103

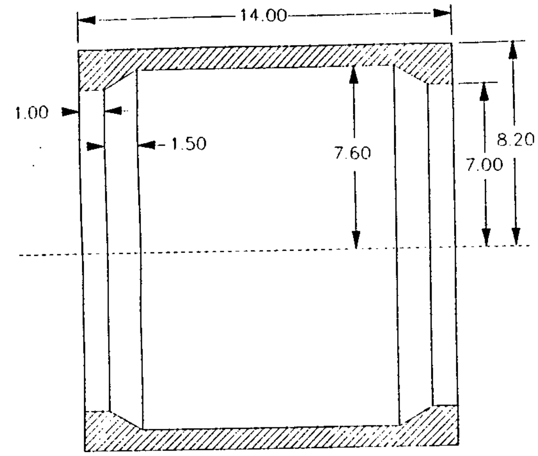

M1 - SHOT 2 - CIRCUMFERENTIAL STRAIN
ANALYTICAL VS. EXPERIMENTAL
MICROSTRAIN
750.
500.
250.
0.
-250.
-500.
-750.
-1000.
-1250.
-1500.
0.
25.
50.
75.
100.
125.
150.
175.
200.
TIME (MICROSECONDS)
ANALYTICAL
GAGE 104


M1 - SHOT 2 - LONGITUDINAL STRAIN
ANALYTICAL VS. EXPERIMENTAL
MICROSTRAIN
4500.
4000.
3500.
3000.
2500.
2000.
1500.
1000.
500.
0.
-500.
-1000.
-1500.
-2000.
-2500.
-3000.
-3500.
0.
25.
50.
75.
100.
125.
150.
175.
TIME (MICROSECONDS)
ANALYTICAL
REFINED ANALYTICAL
GAGE 103

SCALING OF DYNAMIC TESTS

- CURRENT SMALL SCALE MODEL TEST METHODS WERE DEVELOPED FOR ISOTROPIC METALLICS IN WHICH HOPKINSON SCALING IS APPLICABLE UP TO AND EVEN BEYOND YIELD
- SCALING OF THE CYLINDER DYNAMIC TESTS WILL BE INVESTIGATED WITH 16 IN. DIAMETER MODELS DESIGNATED 2XM
- TWO MODELS HAVE BEEN FABRICATED USING S2/ERL 2258 AND THE SAME MATERIAL ORIENTATION AND FABRICATION METHODS AS MODELS M2 AND M3
- 2XM MODELS WILL BE TESTED USING THE SAME TEST SET-UP AS M2 AND M3 WITH THE STANDOFF AND CHARGE SIZE SCALED UP APPROPRIATELY
- IN ADDITION TO DETERMINING HOW WELL STRUCTURAL RESPONSE SCALES, COMPARISONS OF DAMAGE AND FAILURE MECHANISMS OF THE DIFFERENT SIZE MODELS WILL BE MADE

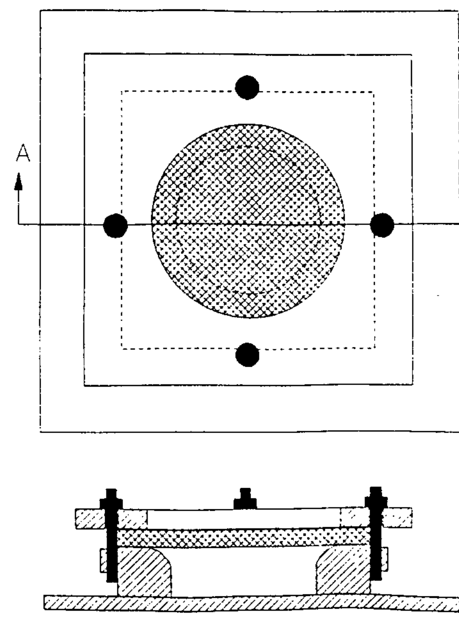
STRUCTURAL CONFIGURATION
2XM MODELS



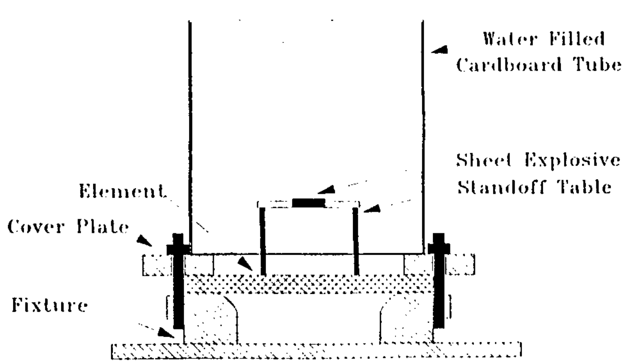
Notes: All Dimensions in Inches

COMPOSITE ELEMENT SHOCK TEST

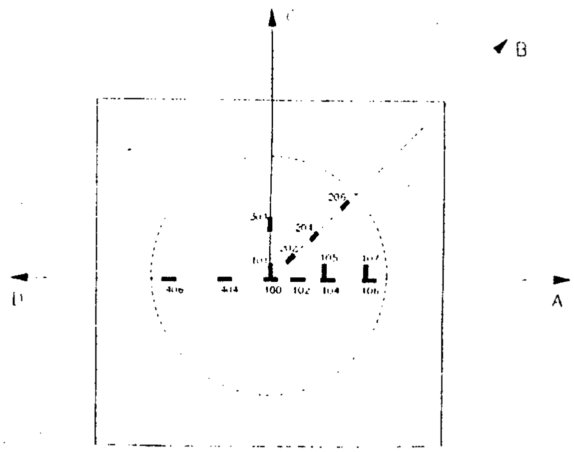
- AXISYMMETRIC PLATE ELEMENT TO INVESTIGATE THICK SECTION RESPONSE
- EXTENSION OF STRUCTURAL TOUGHNESS ELEMENT LESSONS TO COMPOSITE MATERIALS (IMPORTANCE OF STRUCTURE ON MATERIAL FAILURE)
- EXPLOSIVE IN WATER COLUMN TO PROVIDE REALISTIC SHOCK EXCITATION
- THROUGH THICKNESS STRESS WAVE AND FLEXURAL STRUCTURAL RESPONSE INCLUDED
- ELEMENT CAN BE TAILORED TO RECREATE STRAINS PREDICTED FOR FULL SCALE STRUCTURE (BASED ON MODELS OR FEM)

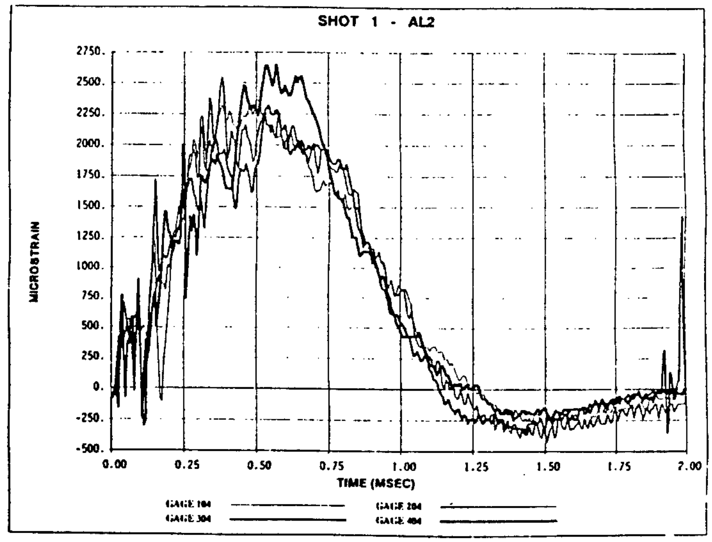


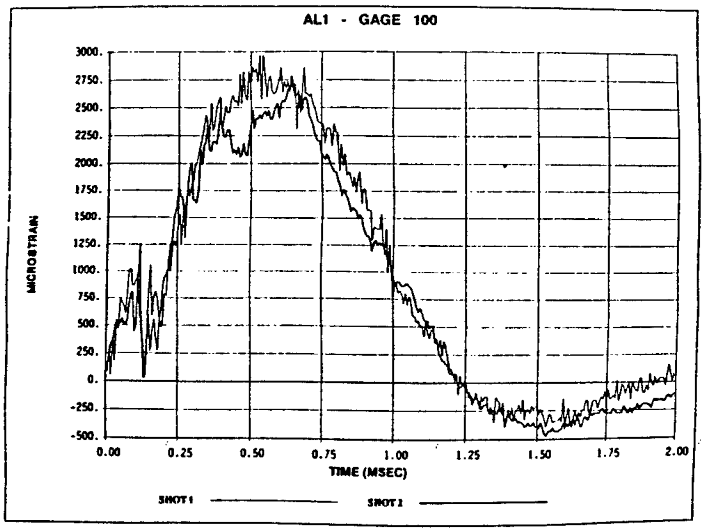
ELEMENT TEST CONFIGURATATION



GAGE LOCATIONS ON CALIBRATION ELEMENT







STATUS

- RELATIVELY SIMPLE, REPEATABLE EXPLOSIVE TEST METHODS HAVE BEEN DEVELOPED FOR BOTH CYLINDRICAL AND FLAT PLATE STRUCTURAL ELEMENTS
- STRUCTURAL LEVEL FINITE ELEMENT ANALYSES OF THE CYLINDRICAL MODELS HAVE REPRODUCED THE LOWER SEVERITY TESTS VERY WELL
- HIGHER SEVERITY TESTS IN WHICH MATERIAL DEGRADATION AND/OR NONLINEARITY OCCUR ARE NOT REPRODUCED WELL ANALYTICALLY
- MODELS HAVE BEEN FABRICATED TO INITIATE THE INVESTIGATION OF SCALING OF COMPOSITE STRUCTURES UNDER THESE LOADING CONDITIONS
- FLAT PLATE ELEMENT TEST HAS BEEN CALIBRATED WITH INSTRUMENTED ALUMINUM PLATES AND STATIC AND DYNAMIC TESTS OF FIRST 1 IN. THICK COMPOSITE ELEMENT IS ONGOING
- STATIC AXISYMMETRIC F.E.M. ANALYSES PERFORMED ON PLATE ELEMENT TO DETERMINE BOUNDARY CONDITIONS - RIGOROUS DYNAMIC ANALYSES WITH POST-PROCESSING APPLICATION OF FAILURE THEORIES WILL ACCOMPANY COMPOSITE ELEMENT TESTS

A MODEL FOR HIGH STRAIN-RATE RESPONSE OF THICK COMPOSITES*

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Mechanics of Materials Branch, Naval Research Laboratory,
Washington, D.C. 20375-5000

A multidimensional constitutive model is developed to study the response of thick composite materials subjected to high rate loading processes. A first order tensor (vector) description of damage is included in a thermodynamic formulation with internal variables for the derivation of the constitutive equations for materials having transverse isotropic symmetry. The vector description of damage is appropriate for materials that undergo brittle damage in the form of thin, planar microcracks that develop with a preferred orientation with respect to the directions of loading. For fiber reinforced laminated composite materials these cracks may be interlaminar, which corresponds to delaminations, or may be intralaminar, which is described as matrix cracking.

The material under consideration is a thick laminated composite material with no fiber reinforcement in the through-thickness direction and a balanced arrangement of fibers in-plane, and is thus considered to be transversely isotropic. The damage is characterized by the vector $\underset{\sim}{V}$ with the V_1 component representing delamination damage and a combination of V_2 and V_3, denoted by V_*, representing matrix damage. Whereas V_1 damage evolves, perhaps catastrophically, to complete separation, the V_* damage evolves to a spatially saturated state of damage due to the presence of fibers in plane. In both cases the damage parameter is taken to vary between zero for the virgin state and one for the completely damaged state.

Considering small strain and axisymmetry, a free energy function is formed containing six invariants of the strain tensor and damage vector. Differentiation of the free energy function yields the usual relationship for the three normal and one shear component of stress and strain. Here, however, the C_{ij}'s are taken to be functions of the damage parameters. The prescribed damage dependence is then placed on the elastic parameters, i.e. elastic modulii, Poisson's ratios, and the shear modulus, as determined by experimental data.

Rate dependence is introduced into the constitutive model through the damage evolution equations, which are taken as a function of the current state of damage and state of stress above a scalar threshold function F. Both the V_1 and V_* types of damage are assumed to be governed by a Mohr-Coulomb type of surface in the space of equivalent normal and shear stresses. Each threshold surface is defined in terms of three parameters - a tension growth threshold, a shear growth threshold, and a tangent parameter. These parameters are each taken to be functions of damage, such that the threshold to produce more damage in the material evolves during the loading process. The rate of damage accumulation is taken to be a function of the distance in stress space from the current state of stress to the surface F, analogous to overstress formulations of viscoplasticity. The principal difference in the evolution equations for the two types of damage is that the equation for V_1 damage contains a singularity as V_1 goes to 1, simulating catastrophic brittle failure. The equation for V_* contains no singularity, however the threshold parameters increase with increasing damage, resulting in a saturation of damage as V_* goes to 1. In addition to the two damage parameters the model utilizes a rate-independent in-plane strain criteria, which defines material failure.

Calculations are performed for two homogeneous loading conditions over a range of strain rates The first is for uniaxial stress in the direction normal to the fibers, and the second is for radial deformation in the fiber plane with zero stress in the through-thickness direction. For loading in the through thickness direction at the lowest rate, the response is indicative of a perfectly brittle material with essentially instantaneous damage accumulation. At higher rates, the time required for damage accumulation permits higher stress levels to be reached prior to a more gradual softening behavior. For loading in the plane of the fibers, low rate deformation results in a rapid maturation of damage, producing a bilinear response. At intermediate rates, damage matures gradually, permitting some overstress to occur, followed by gradual softening. At the highest rate the damage has insufficient time to develop prior to reaching the strain failure criteria, resulting in low amounts of damage.

* This work supported by DARPA Naval Technology Office.

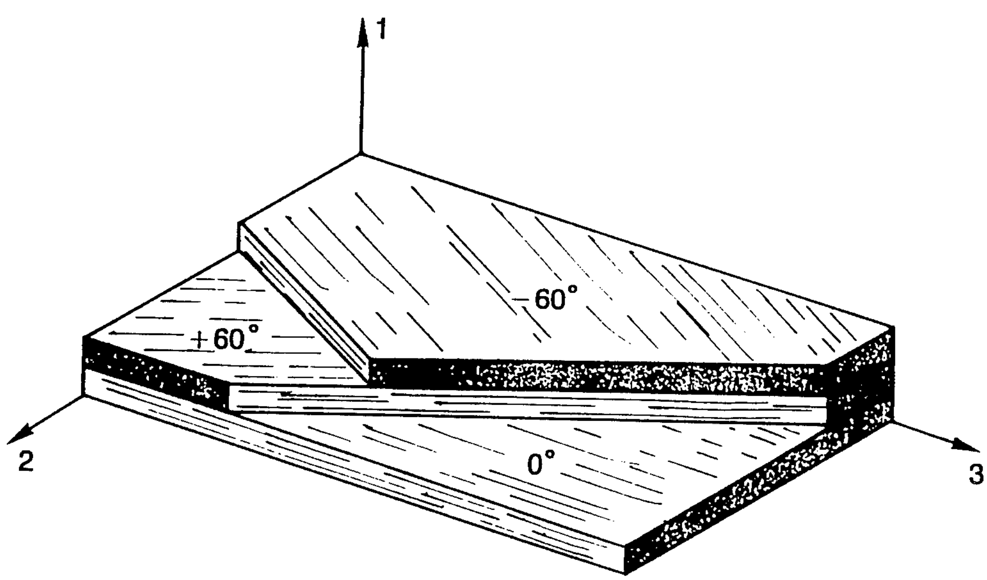
OBJECTIVE:

- MATERIAL RESPONSE PREDICTIONS IN THICK COMPOSITES UNDER HIGH LOADING RATE
- INCLUDE PROGRESSIVE CONTINUUM MATERIAL DAMAGE FOR SOFTENING EFFECTS ON STRESS WAVE RESPONSE AND FOR PREDICTIONS OF END DAMAGE STATE

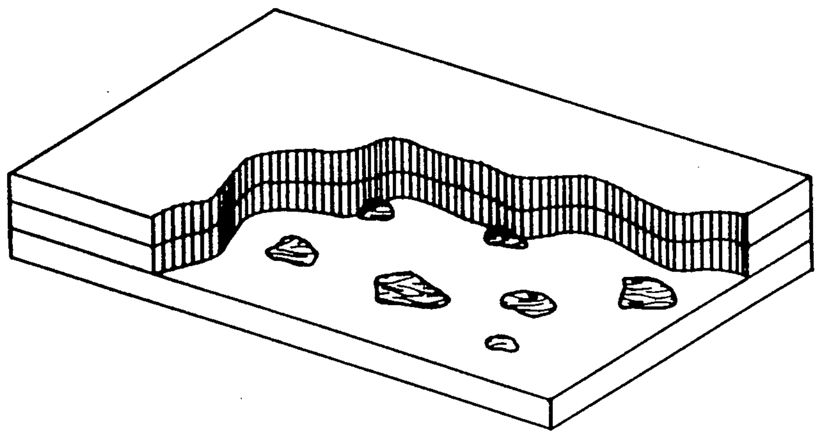
APPROACH:

- ENERGY BASED DEVELOPMENT WITH INTERNAL VARIABLES FOR DAMAGE

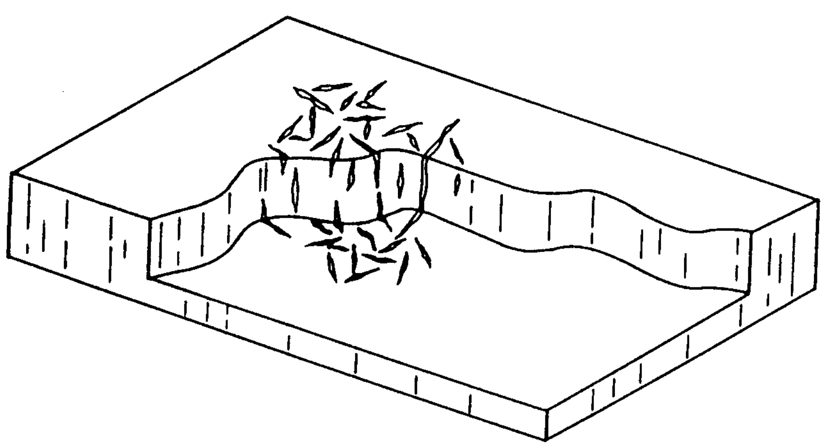
QUASI-TRANSVERSELY ISOTROPIC MATERIAL LAYUP
(ALLOWS APPROXIMATION OF 2-D AXISYMMETRIC
MATERIAL RESPONSE AND DAMAGE DEVELOPMENT)



SPALLATION (DELAMINATION) DAMAGE
CHARACTERIZED BY V_1 DAMAGE VARIABLE



MATRIX CRACKING DAMAGE (NOT INCLUDING FIBER BREAKAGE)
CHARACTERIZED BY $V_* = \sqrt{V_2^2 + V_3^2}$ DAMAGE VARIABLE



FORMULATION OF CONSTITUTIVE EQUATIONS FOR DAMAGING MATERIAL

POSTULATE A FREE ENERGY FUNCTION UNDER CONDITIONS OF TRANSVERSE ISOTROPY AND AXISYMMETRY WITH RESPECT TO THE 1- AXIS (through thickness)

$$\rho\psi = f(I_1, I_2, \ldots, I_6)$$

WITH STRAIN AND DAMAGE INVARIANTS

$$I_1 = \varepsilon_{11} \qquad I_4 = \varepsilon_{12}^2$$

$$I_2 = \varepsilon_{22} + \varepsilon_{33} \qquad I_5 = V_1^2$$

$$I_3 = \varepsilon_{22}^2 + \varepsilon_{33}^2 \qquad I_6 = V_*^2 = V_2^2 + V_3^2$$

STRESS-STRAIN CONSTITUTIVE RELATIONS DERIVE FROM THE FREE ENERGY FUNCTION

$$\sigma_{ij} = \partial(\rho\psi)/\partial\varepsilon_{ij}$$

FOR SMALL STRAINS (linear with deformation, nonlinear with damage), THIS TAKES THE USUAL TRANSVERSELY ISOTROPIC FORM

$$\begin{pmatrix} \sigma_{11} \\ \sigma_{22} \\ \sigma_{33} \\ \sigma_{12} \end{pmatrix} = \begin{bmatrix} C_{11} & C_{12} & C_{12} & 0 \\ C_{12} & C_{22} & C_{23} & 0 \\ C_{12} & C_{23} & C_{22} & 0 \\ 0 & 0 & 0 & C_{44} \end{bmatrix} \begin{pmatrix} \varepsilon_{11} \\ \varepsilon_{22} \\ \varepsilon_{33} \\ \varepsilon_{12} \end{pmatrix}$$

WHERE THE C_{ij} ARE FUNCTIONS OF DAMAGE AND, FOR THE AXISYMMETRIC CASE, TAKE

$\sigma_{11} = \sigma_z$, $\sigma_{22} = \sigma_r$, $\sigma_{33} = \sigma_\Theta$, $\sigma_{12} = \sigma_{zr}$, $\sigma_{13} = \sigma_{23} = 0$

RELATIONSHIP OF CONSTITUTIVE COEFFICIENTS TO ELASTIC CONSTANTS

$$C_{11} = (1 - \nu_{23})E_{11}/(1 - \nu_{23} - 2\nu_{12}^2E_{22}/E_{11})$$

$$C_{12} = \nu_{23}E_{22}/(1 - \nu_{23} - 2\nu_{12}^2E_{22}/E_{11})$$

$$C_{22} = \frac{1}{1 + \nu_{23}}(1-\nu_{12}^2E_{22}/E_{11})E_{22}/(1-\nu_{23}-2\nu_{12}^2E_{22}/E_{11})$$

$$C_{23} = \frac{1}{1 + \nu_{23}}(\nu_{23} + \nu_{12}^2E_{22}/E_{11})E_{22}/(1-\nu_{23}-2\nu_{12}^2E_{22}/E_{11})$$

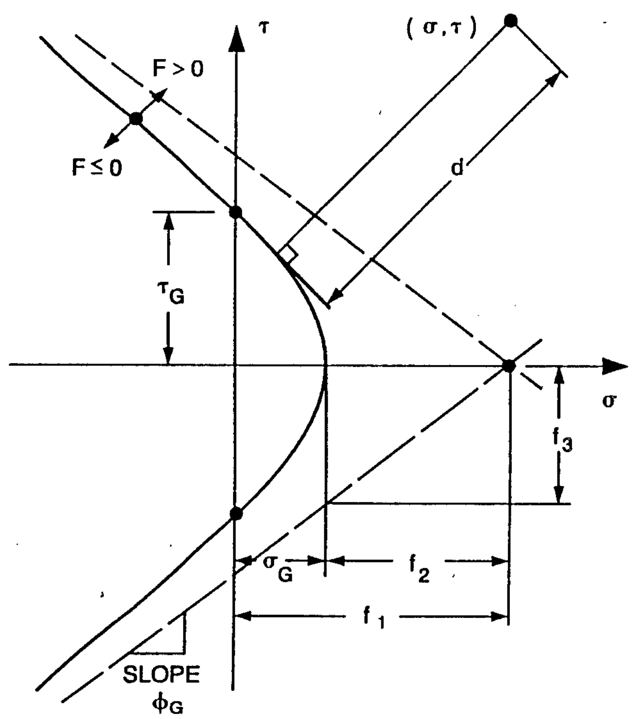
$$C_{44} = 2G_{12}$$

PRELIMINARY DAMAGE (SOFTENING) EFFECTS IMPOSED ON THE ELASTIC CONSTANTS

$$E_{11} = (1 - V_1^2)E_{11}^{\circ}\ ,\ E_{22} = (1-\alpha_1V_*^2)E_{22}^{\circ}\ ,\ \nu_{23} = (1-\alpha_4V_*^2)\nu_{23}^{\circ}$$

$$G_{12} = (1 - V_1^2)(1 - \alpha_2V_*^2)G_{12}^{\circ}\ ,\ \nu_{12} = (1-V_1^2)(1-\alpha_3V_*^2)\nu_{12}^{\circ}$$

THRESHOLD SURFACE FOR THE ONSET OF DAMAGE



THE DAMAGE CONSTITUTIVE RELATIONS
(THRESHOLD AND EVOLUTION)

ASSUME A Mohr-Coulomb THRESHOLD SURFACE DEFINED BY **F** = 0 WHERE

$$F = \sqrt{1 + (\tau/f_3)^2} - (f_1 - \sigma)/f_2$$

THIS FORM WILL CONTROL THE ONSET OF BOTH V_1 and V_* DAMAGE WHERE STRESS THRESHOLDS σ_G AND τ_G ARE RELATED TO THE PARAMETERS f_i BY

$$\sigma_G = f_1 - f_2$$

$$\tau_G = f_3 \sqrt{(f_1/f_2)^2 - 1}$$

$$\phi_G = f_3/f_2$$ (A COULOMB FRICTION TANGENT)

EVOLUTION EQUATIONS FOR V_1- DAMAGE (SPALLATION AND DELAMINATION)

TAKE $\sigma = \sigma_{11}$ AND $\tau = \sigma_{12}$ IN **F**

RELATE SPECIFIC THRESHOLDS TO DAMAGE BY

$$\sigma_G=(1 - V_1^2)\sigma_{G0},\ \tau_G = (1 - V_1^2)\tau_{G0},\ \phi_G=\phi_{G0}+V_1^2(\phi_{G1}-\phi_{G0})$$

WHERE σ_{G0}, τ_{G0} AND ϕ_{G0} ARE VIRGIN VALUES AND ϕ_{G1} IS AND END STATE. THEN TAKE $d_1 = d = 0$ IF $\mathbf{F} \leq 0$

= CALCULATED DISTANCE IN σ,τ- SPACE TO RESPONSE POINT IF $\mathbf{F}>0$

EVOLUTION EQUATION

$$\dot{V}_1 = (d_1/\sigma_{G0})^{n_1}/[\eta_1(1-V_1^2)]$$

WITH η_1 - RATE CONSTANT n_1 - EXPON ENT IN POWER LAW

$1/(1-V_1^2)$ FACTOR CAUSES $\dot{V}_1 \rightarrow +\infty$ AS $V_1 \rightarrow 1$

EVOLUTION EQUATION FOR V_* DAMAGE (IN-PLANE MATRIX CRACKING)

TAKE $\sigma=\frac{1}{2}(\sigma_{22}+\sigma_{33})$, $\tau=\sqrt{\sigma_{12}^2+\frac{1}{4}(\sigma_{22}-\sigma_{33})^2}$,

WITH SPECIFIC THRESHOLDS GIVEN BY

$$\bar{\sigma}_G=\bar{\sigma}_{G0}/(1-V_*^2),\ \bar{\tau}_G=\bar{\tau}_{G0}/(1-V_*^2),\ \bar{\phi}_G=\bar{\phi}_{G0}+V_*^2(\bar{\phi}_{G1}-\bar{\phi}_{G0})$$

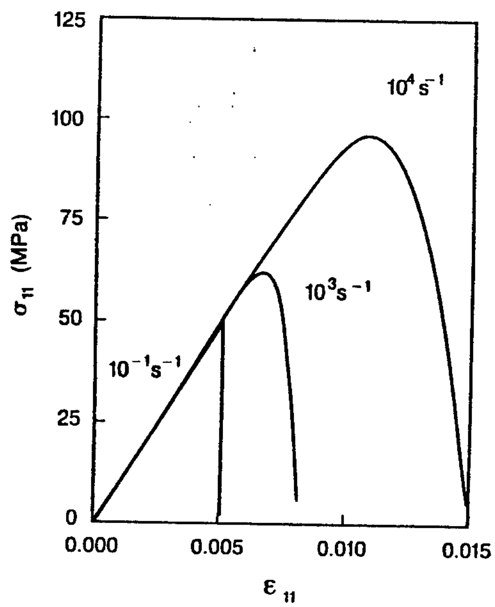
THE $1/(1-V_*^2)$ FACTORS CAUSE SATURATION OF V_* DAMAGE

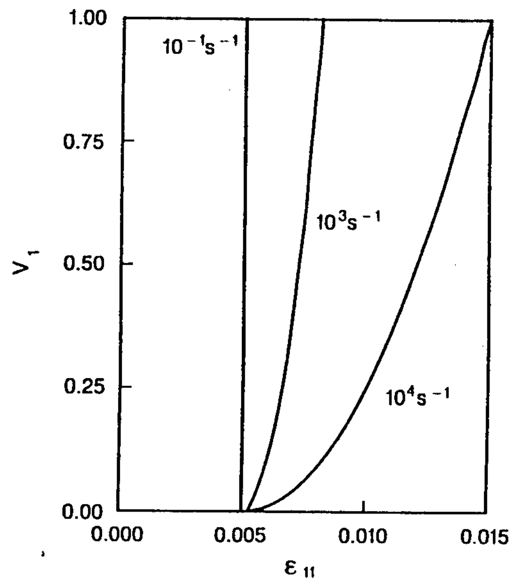
EVOLUTION EQUATION

$$\dot{V}_* = (d_*/\bar{\sigma}_{G0})^{n_*}/\eta_*$$

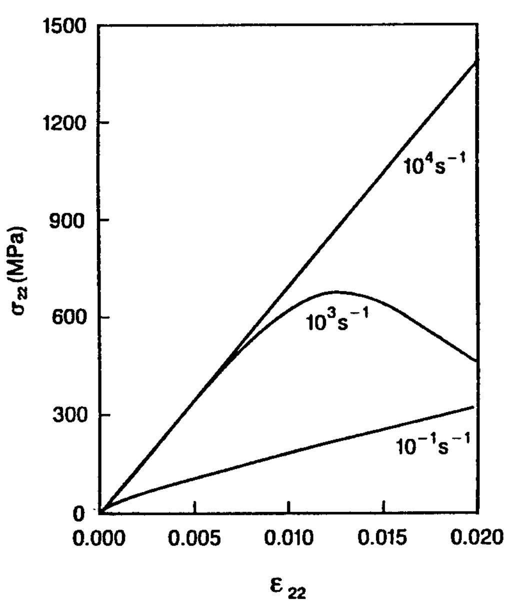
WITH MATERIAL CONSTANTS η_* AND n_* AND WITH $d=d_*$ AGAIN BEING THE DISTANCE FROM THE THRESHOLD SURFACE.

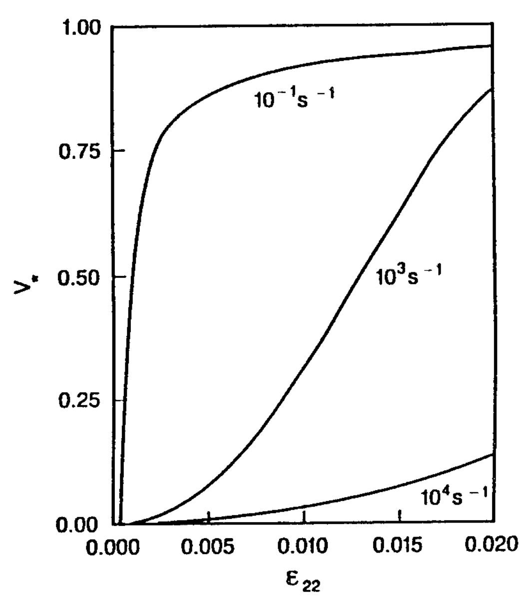
STRESS AND DAMAGE RESPONSE UNDER UNIAXIAL STRESS ($\sigma_{22} = \sigma_{33} = 0$) IN THROUGH-THICKNESS DIRECTION AT VARIOUS STRAIN RATES





STRESS AND DAMAGE RESPONSE UNDER BIAXIAL STRESS ($\sigma_{11} = 0$) IN-PLANE AT VARIOUS STRAIN RATES





Experimental Measurements of Crack Tip Deformations at Interface in Composites *

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ABSTRACT

The state of stress and strain in the neighborhood of a crack situated along the interface of a bimaterial is almost always three dimensional. However, most treatments of the problems are two-dimensional in nature. In order to shed some light on the three dimensionality of the interfacial crack problem we developed an embedded speckle method whereby interior deformation of a transparent specimen can be measured. The bimaterial specimen is made of two types of polyester raisns with Young's modulii being 3000 Mpa and 229.2 Mpa and Poisson's ratio being 0.34 and 0.42, respectively. The interface is created by casting one of the rasins onto a cured block of the other rasin. A 20 μm thick Teflon sheet is placed along part of the bonding surface to simulate an interfacial crack. The specimen is in the form of a beam with the dimension 38.1 $\times$38 $\times$ 146 mm and the interface is at the half length of the beam with the crack extended into a depth of 18.6 mm from one edge. The specimen consists of four identical slices of 9.5 mm thick and is glued together using the softer polyester rasins. The gluing layers are about 25 μm thick and they contain a small amount of 10 μm size glass beads which serve as the speckle patterns. A thin layer of glass beads is also sprayed onto the front surface of the beam. It is then placed into a index matching fluid and loaded in four-point-bend. Before and after the deformation the speckle patterns on the front surface together with those at the 1/4 and 1/2 thickness planes are recorded. Specklegrams are formed by pairing the two speckle patterns, before and after deformation, at the corresponding planes. Optical Fourier Processing of the specklegrams yield displacement information at three planes in the form of Young's fringes. In particular we probe the displacement distribution of the two crack faces. Absolute displacement of the crack faces are obtained and displacement jumps are deduced by taking the difference. It is found that the opening displacement jump at the surface is much larger than that of the interior planes. This implies that the stress intensity along the crack front is not uniform. It is higher at the surface. This is opposite to the case of a homogeneous beam. Thus, should the crack propagate, the

crack front has a shape similar to an inverse thumb nail. That is the crack front at the surface would advance faster than the interior. Numerical results by R. Barsoum also predicated such a phenomenon. Indeed the experimentally obtained singularity index is also fairly close to that obtained numerically by Barsoum.

* Prepared for presentation at DoD/NASA Mechanics of Composites Review, 24-25 Oct. 1990, Dayton, Ohio.

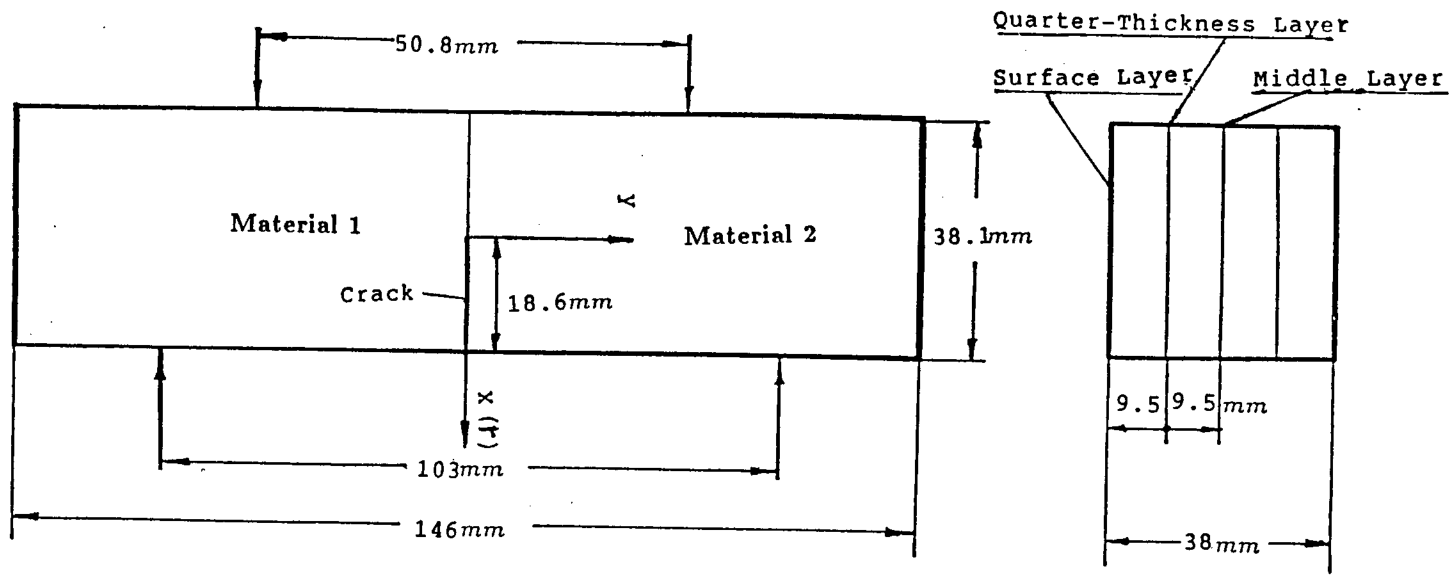


Fig.1 Bimaterial Specimen.

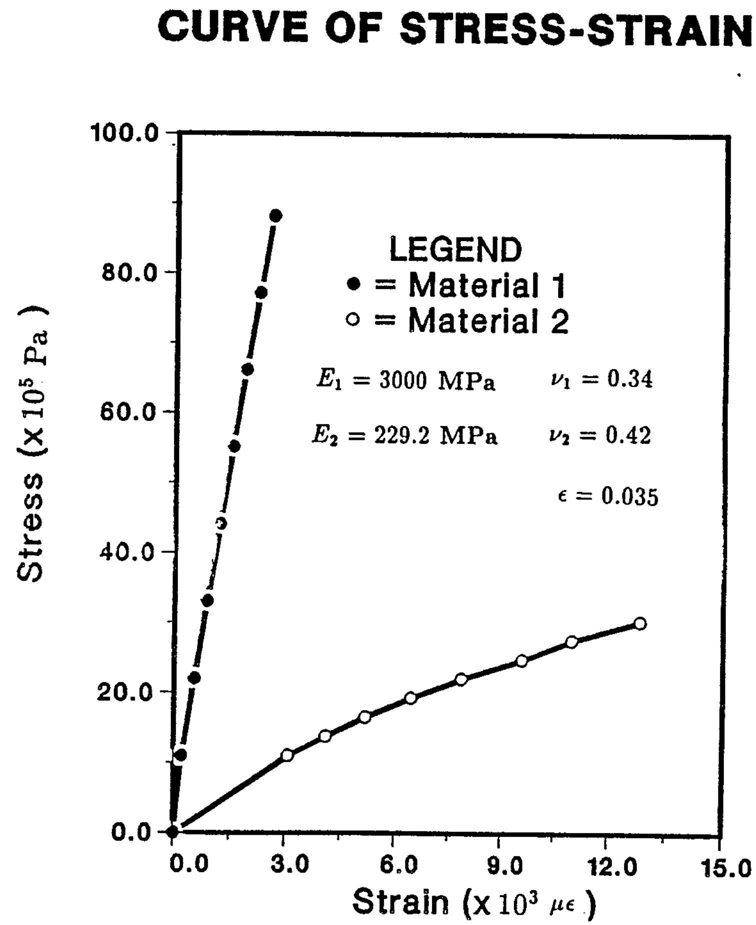


Fig. 2 Material Properties.

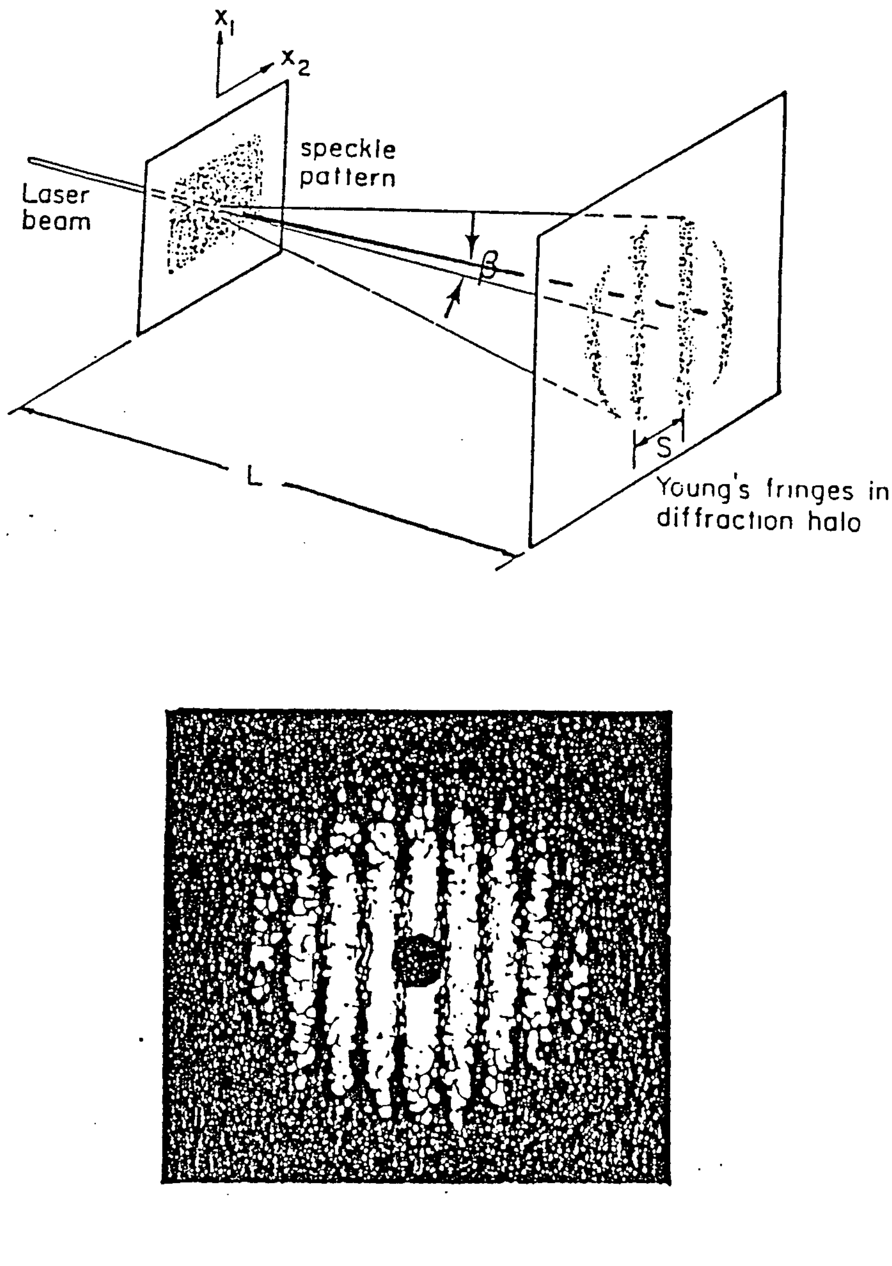
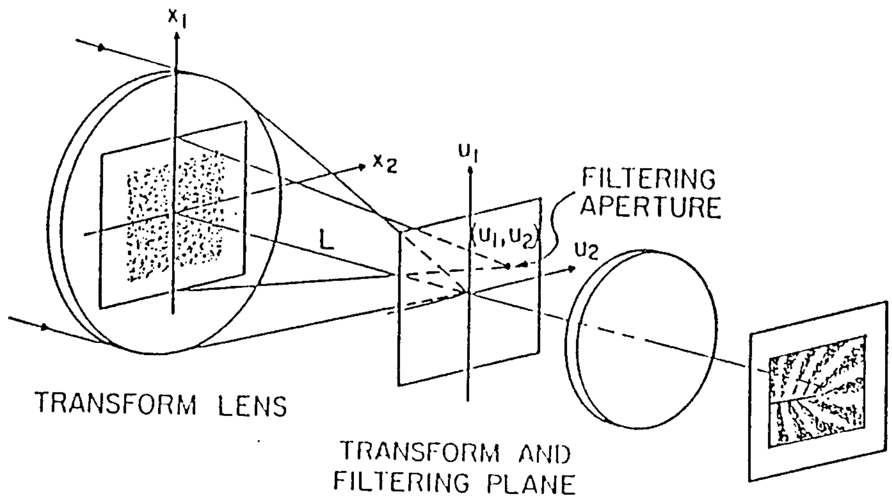


Fig. 3a Pointwise Fourier processing resulting in Young's fringes representing the displacement vector at the point of probing.



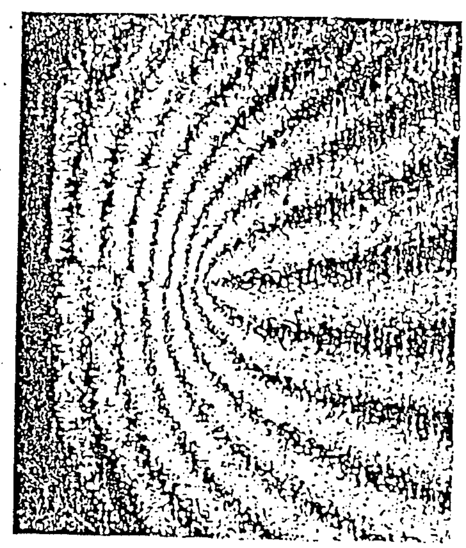
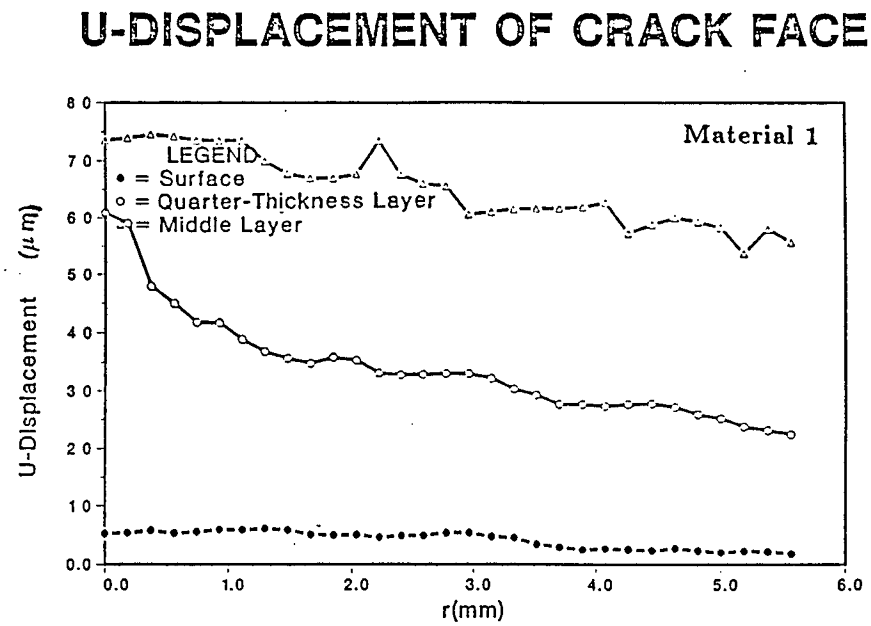


Fig. 3b Full-field Fourier processing resulting in isothetics -- contours of displacement compoment.



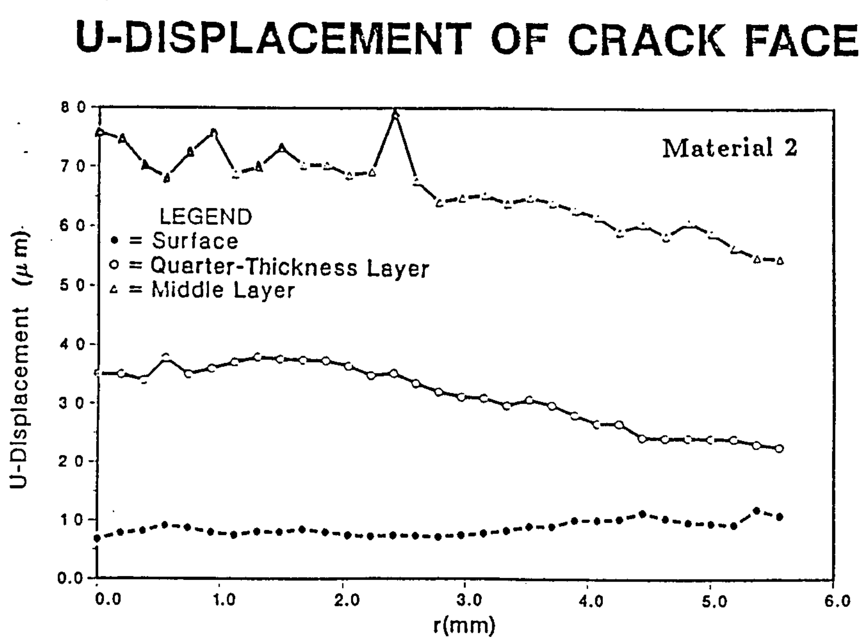
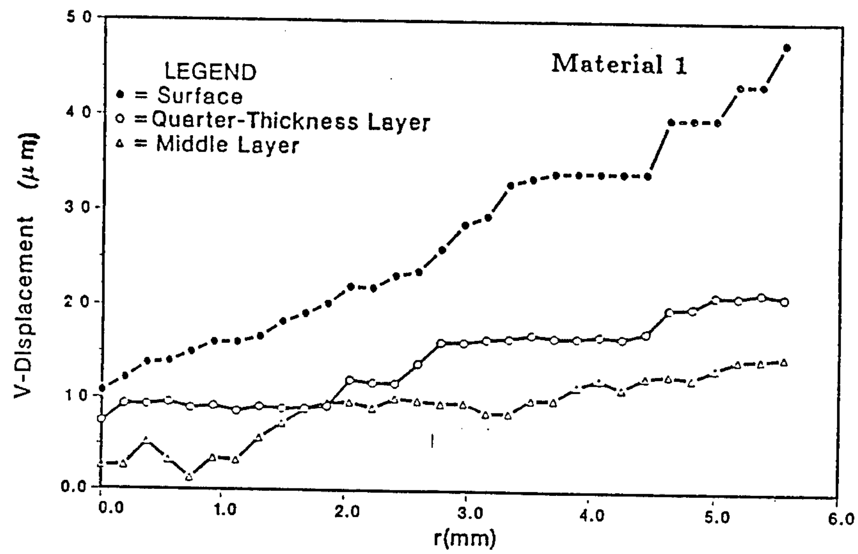


Fig.4(a) Displacement u of the Crack Faces.

V-DISPLACEMENT OF CRACK FACE



V-DISPLACEMENT OF CRACK FACE

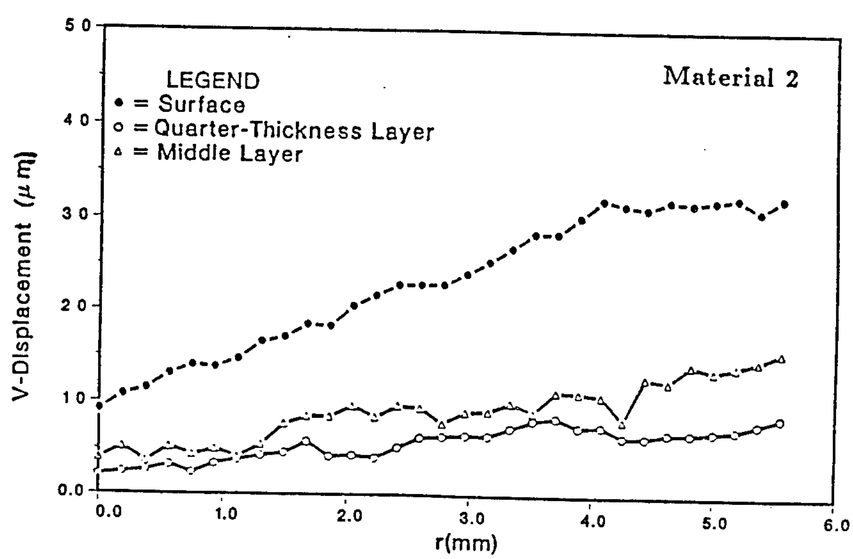


Fig.4(b) Displacement v of the Crack Faces.

ΔV-DISPLACEMENT OF CRACK FACE

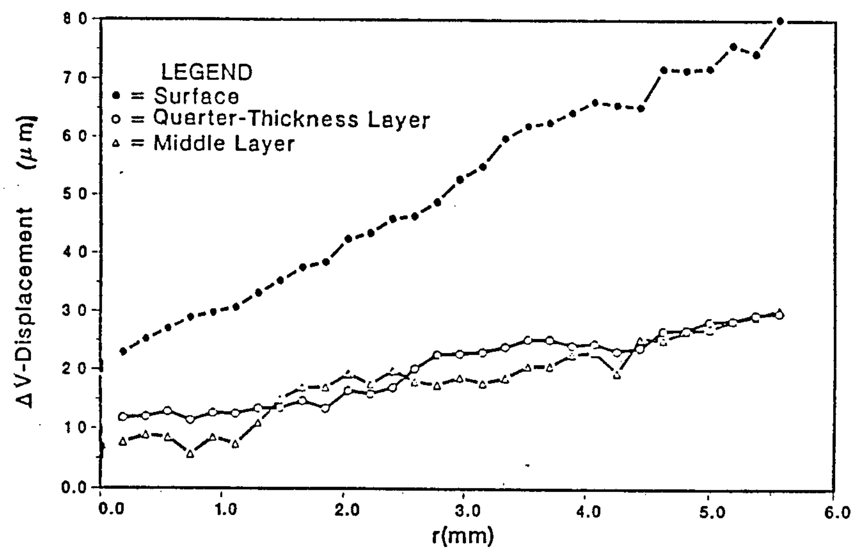


Fig. 5 (a)   Opening Displacement Δv.

ΔU-DISPLACEMENT OF CRACK FACE

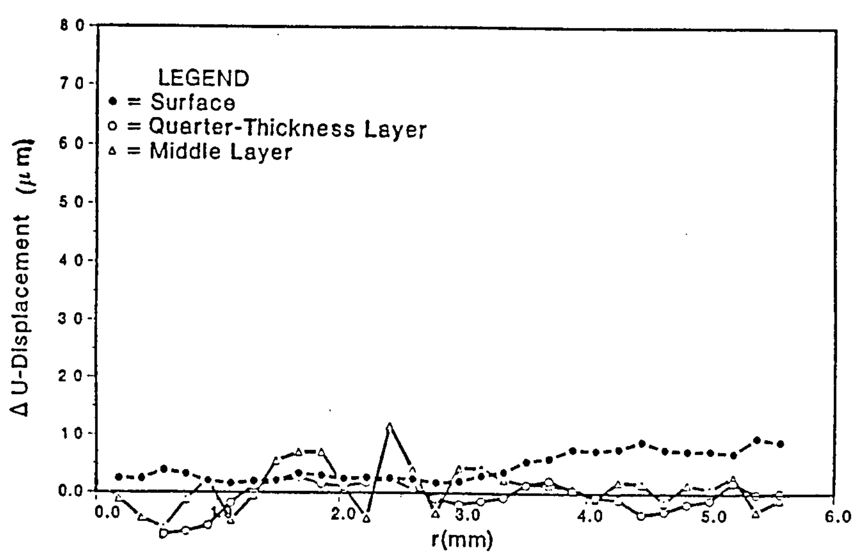


Fig. 5 (b)   Sliding Displacement Δu.

∇ V-DISPLACEMENT OF CRACK FACE

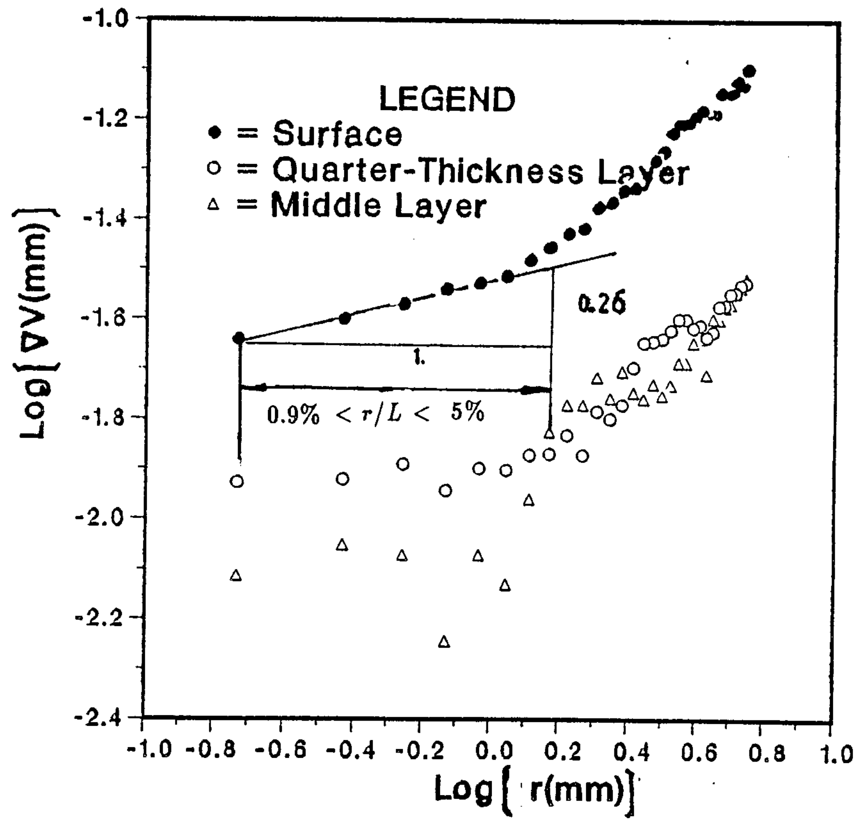


Fig.6 Curve of log Δv *vs* log r.

COMPRESSIVE PROPERTIES OF ADVANCED COMPOSITES FROM A NOVEL SANDWICH TEST SPECIMEN

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ABSTRACT

The intrinsic compression strength of advanced composites is one of the more difficult properties to measure as the results often depend on the loading geometry and test conditions. Slight variations in specimen geometry can result in an eccentricity of the applied load which may lead to premature failure via specimen buckling. Consequently, several test methods have been developed to attempt accurate measurements of this property. Although a number of these tests have been incorporated into an ASTM standard, results from different groups on the same composite system and by the same technique vary widely, questioning the reliability of the numbers reported. To obtain consistent, meaningful data, premature failure must be avoided, and these techniques and the corresponding test fixtures have been modified, in an attempt to accomplish this. Alternatively, the test specimen itself can be redesigned to ensure consistent, acceptable failure modes and realistic strengths for any composite system. This is the approach we have elected to take.

Our specimen is a novel symmetric mini-sandwich beam, which is an adaptation of the traditional sandwich specimen. The original sandwich specimen is large, expensive to fabricate, and requires a large volume of composite for a single specimen. Furthermore, Poisson's expansions of composite skin and honeycomb core are not matched, resulting in separation of skin and core in the test. In addition under flexural loading, premature failure frequently occurs from a crushing of the composite skin under the loading pins. The modified mini-sandwich beam overcomes these deficiencies. In our specimen, the honeycomb core is replaced by a slab of neat resin of the same material as the matrix of the composite to be tested. An equal number of prepreg plies are placed on either side of the partially cured resin core and the assembly is co-cured (or co-consolidated) to give a symmetric beam. Simultaneous consolidation of skin and core provides excellent bonding between these components eliminating the need for an intermediate adhesive layer. These modifications at once minimize the difference in Poisson's expansion between skin and core and provide a simple yet inexpensive means of fabricating sandwich panels.

Initial experiments were conducted with unidirectional composite skins, 2 to 4 plies thick. Mini-sandwich specimens of carbon fibers (AS4, IM8, and P-75) and S-glass fibers with an epoxy matrix, and AS4 with a PEEK matrix, were tested in axial compression with the IITRI fixture and also in four-point flexure. Failure occurred predominantly in the gage section at composite stresses and strains substantially higher than observed in corresponding tests on all-composite coupons. The fracture path in the AS4/epoxy specimens was at approximately 75 degrees to the specimen axis, and SEM examinations of the fracture surfaces revealed fiber failure by buckling and also in shear. Failure analyses of these specimens loaded in axial compression suggested initial compressive failure occurring on the outer surface of a skin. This produced an eccentricity in the applied stress promoting instantaneous buckling and catastrophic global failure. With an increase in skin thickness, higher compressive loads were required to promote failure, and with a six-ply skin, gross specimen bending was discerned from back-to-back strain gages. For the core thickness employed in this study therefore (3.18 mm), composite skins were limited to 4 unidirectional plies or less.

With the success in obtaining acceptable failure modes and reproducible, reliable estimates of compression strength, this test specimen was employed to investigate the compressive strengths of multi-directional laminates. Sandwich panels with AS4/3501-6 symmetric 0/90/90/0 laminate skins were fabricated with an epoxy core cured at room temperature to minimize residual stresses. The laminate skin compressive strength was substantially higher than that of corresponding cross-ply laminates. This specimen is also being utilized to investigate the influence of matrix properties on composite compression strength. Epoxy matrix (3501-6) shear modulus doubles over the temperature range of -75F to 300F. Testing sandwich specimens at different temperatures in this range will therefore afford a means of investigating the singular influence of matrix modulus on composite compression strength, while holding all other composite variables constant. In all our investigations, composite failure modes and stresses were compared with those of similar all-composite coupons tested under identical conditions. In this presentation the development of the mini-sandwich specimen, its fabrication, and the results of our tests are reported.

COMPRESSIVE PROPERTIES OF ADVANCED COMPOSITES FROM A NOVEL SANDWICH TEST SPECIMEN

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and

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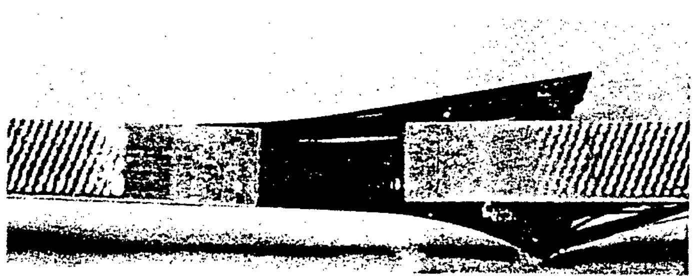
Presented at the Fifteenth Annual Mechanics of Composites Review,
October 25, 1990

CONVENTIONAL TECHNIQUES FOR MEASURING COMPOSITE COMPRESSION STRENGTH

- **Celanese method**
- **IITRI technique**
- **Modified ASTM D695**
- **Sandwich specimens**



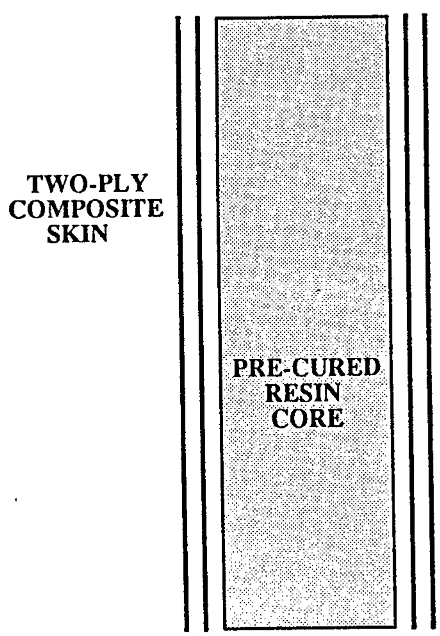
24-PLY UNIDIRECTIONAL AS4/3501-6

Under compressive loading, stress concentrations beneath the tab lead to premature failure via intralaminar splitting. (Specimen width : 6.35 mm)

NOVEL MINI-SANDWICH TEST SPECIMEN

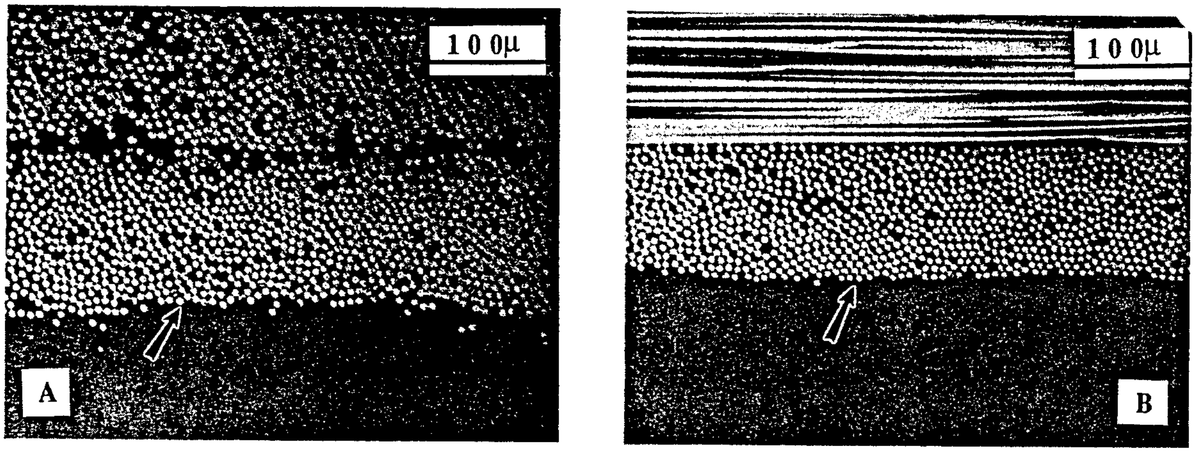
- Requires a relatively small amount of material.
- Simple and inexpensive to fabricate.
- Poisson's expansions of core and skin are better matched than in conventional sandwich specimens
- Can be tested in four-point flexure *and* direct axial compression.
- Specimen dimensions are similar to those for conventional composite coupons.
- Provides greater resistance to buckling in axial compression than composite coupons.
- Provides higher compression strengths than composite coupons with the same test method.

MINI-SANDWICH COMPOSITE PANEL

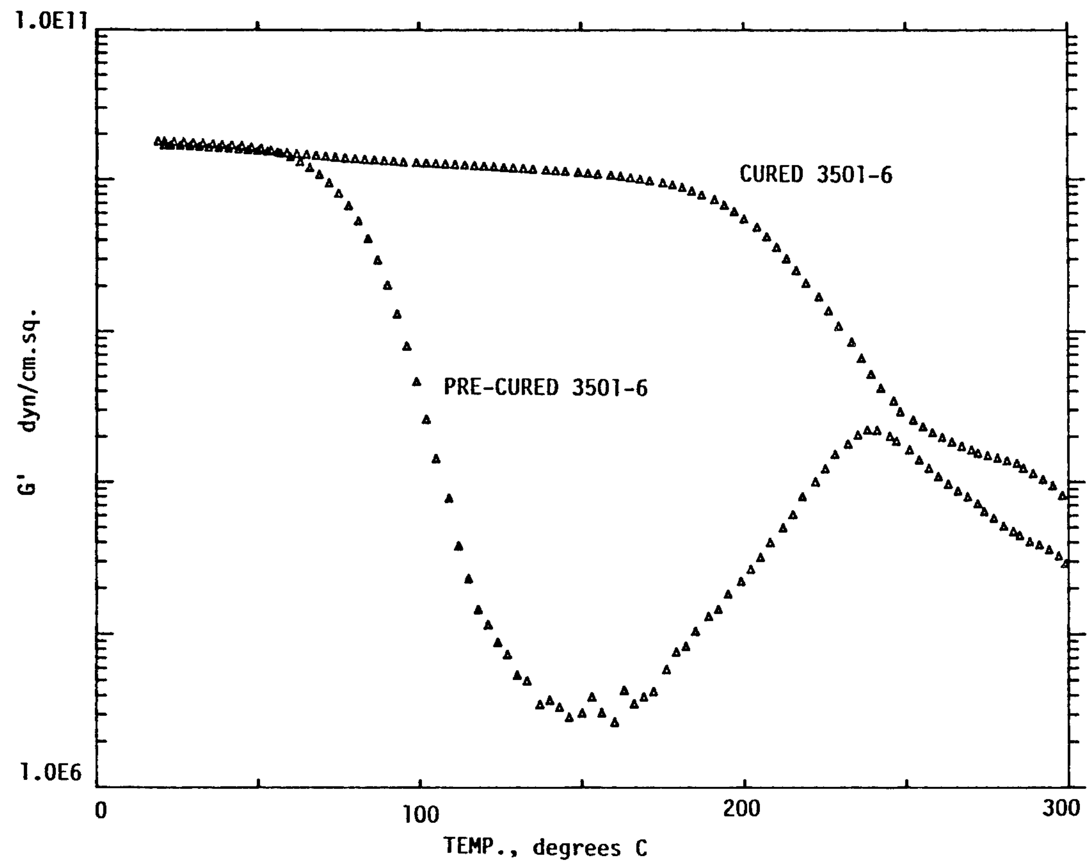


COMPOSITE SYSTEMS STUDIED

CORE	SKIN
EPOXY (3501-6)	AS4/3501-6
	IM8/3501-6
	P-75/934
	S-Glass/1034
	K-149/3501-6
PEEK	AS4/APC-2



Cross-sections of sandwich panels with (A) ($0_2/\pm30$) AS4/APC-2 skin and a PEEK core, and (B) ($0/90_2/0$) AS4/3501-6 skin and an epoxy core, showing excellent adhesion between skin and core. (Arrows indicate the skin/core interface region).



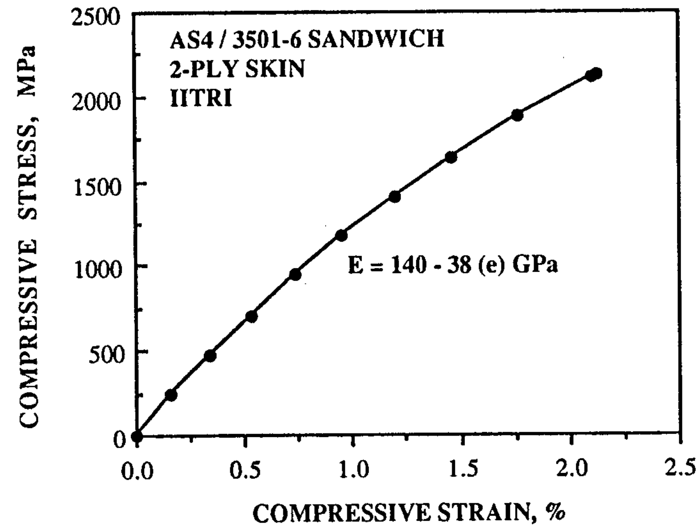
Dynamic shear modulus of a partially cured (4h @ 120C) neat epoxy (3501-6) core prior to sandwich fabrication, and a fully cured casting, as a function of temperature.

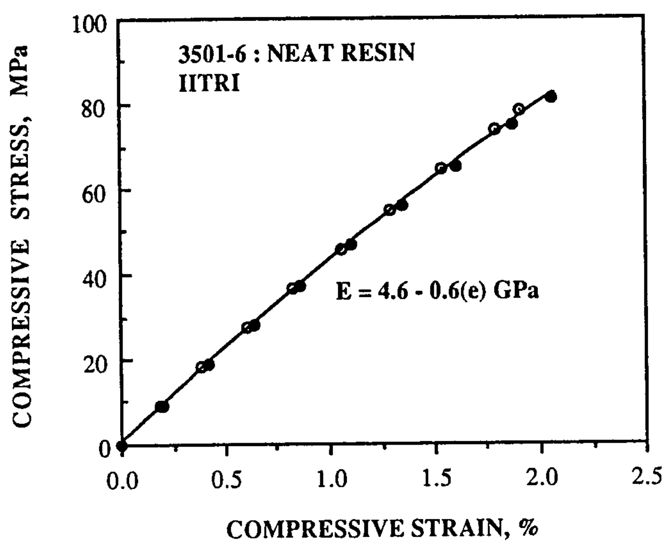
COMPOSITE COMPRESSIVE STRESS IN A MINI-SANDWICH UNDER AXIAL COMPRESSION

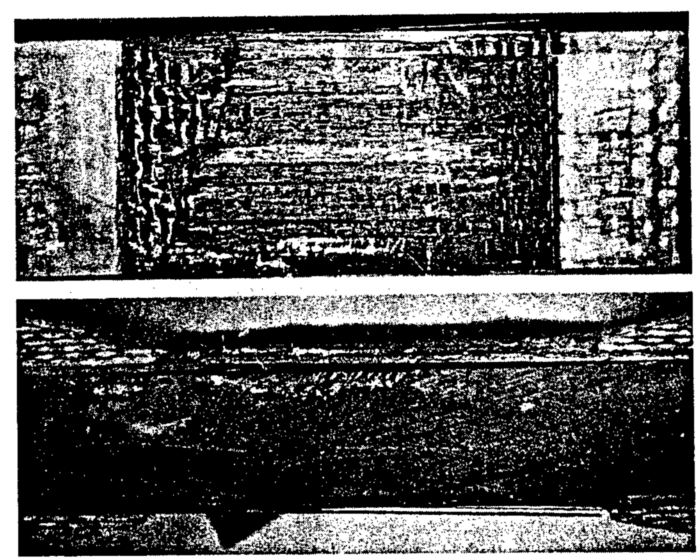
$$E_t = E_s V_s + E_c V_c$$

$$\sigma_s = (\sigma_t - \varepsilon_t E_c V_c) / V_s$$

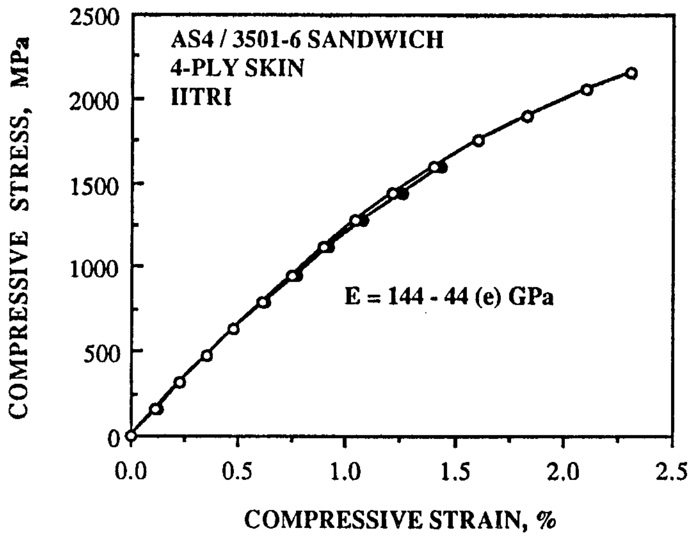
where : t is the total specimen
c is the core (resin)
s is the skin (composite)



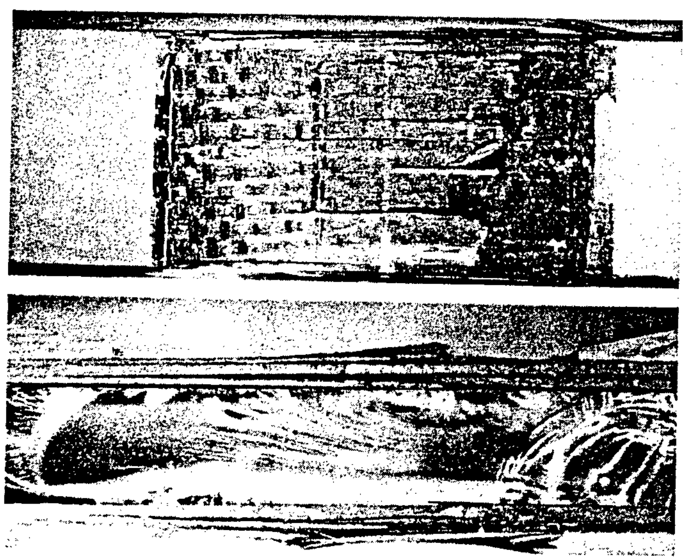




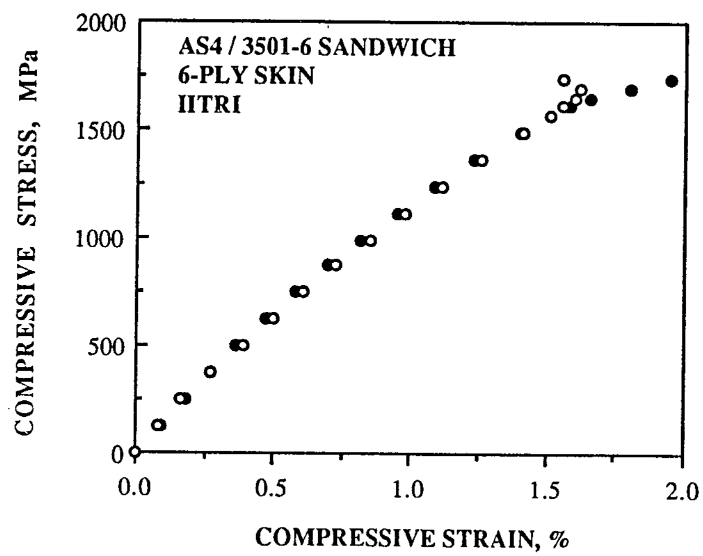
Flat and side views of the gage section of a 2-ply unidirectional AS4/3501-6 sandwich specimen failed in axial compression. (Specimen width : 6.35 mm).

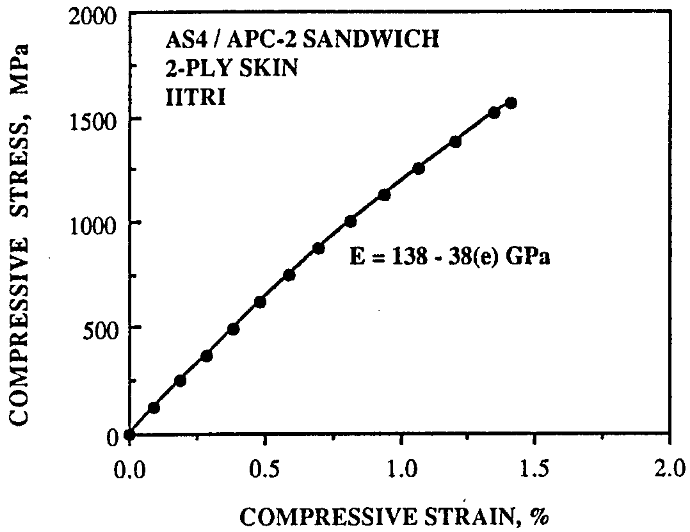


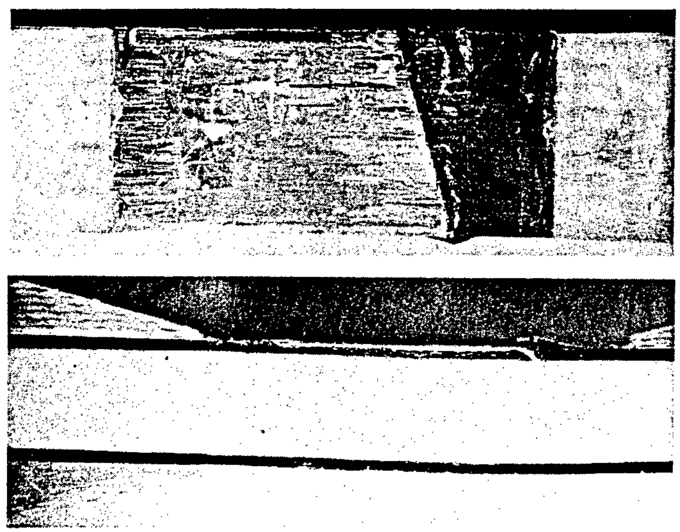
Fracture surface of the 4-ply unidirectional composite skin of an AS4/3501-6 sandwich specimen failed in axial compression, showing shear failure of the fibers.



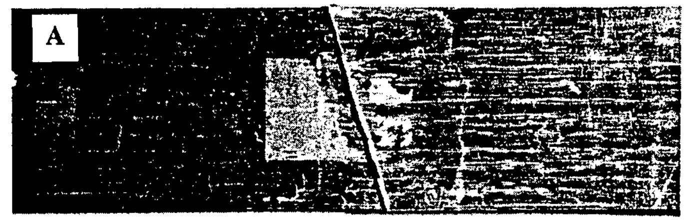
Flat and side views of the gage section of a 4-ply unidirectional AS4/3501-6 sandwich specimen failed in axial compression. (Specimen width : 6.35 mm).

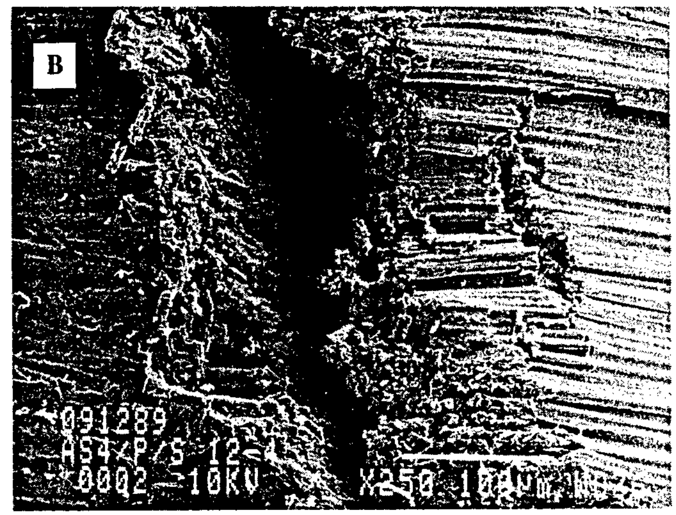






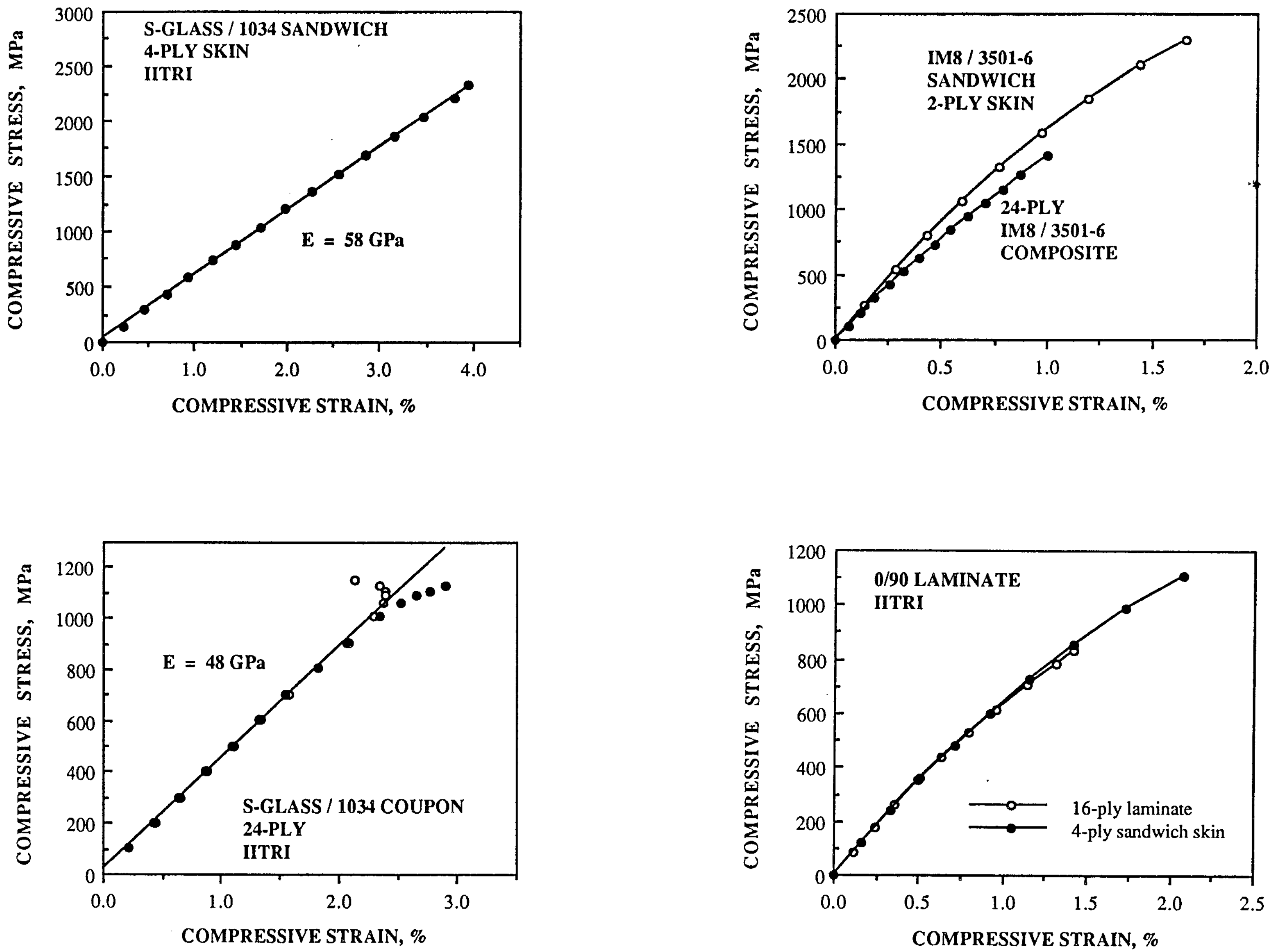
Flat and side views of the gage section of a 2-ply unidirectional AS4/APC-2 sandwich specimen failed in axial compression. (Specimen width : 6.35 mm).





(A) Flat view of the 2-ply unidirectional skin of an AS4/APC-2 sandwich specimen failed in compression under four-point flexural loading. (Specimen width : 12.7 mm)

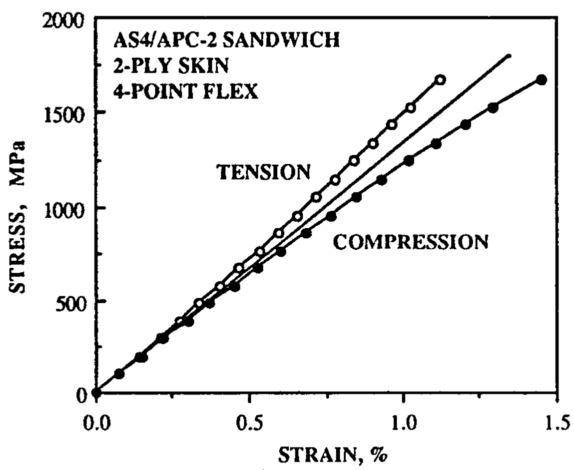
(B) Magnified view of the fracture surface in (A) showing shear failure of the composite.

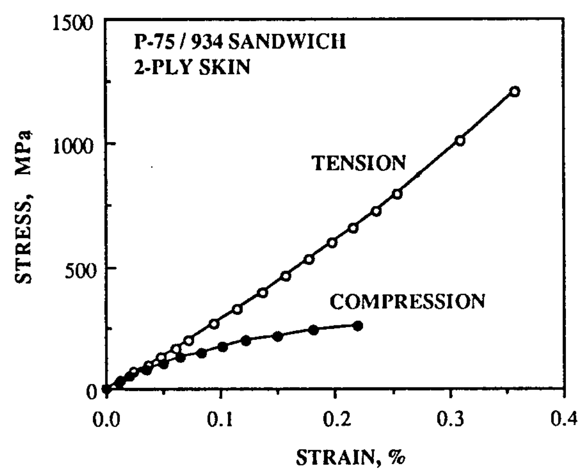

S-GLASS / 1034 SANDWICH
4-PLY SKIN
IITRI
E = 58 GPa
COMPRESSIVE STRESS, MPa
COMPRESSIVE STRAIN, %
IM8 / 3501-6
SANDWICH
2-PLY SKIN
24-PLY
IM8 / 3501-6
COMPOSITE
COMPRESSIVE STRESS, MPa
COMPRESSIVE STRAIN, %
E = 48 GPa
S-GLASS / 1034 COUPON
24-PLY
IITRI
COMPRESSIVE STRESS, MPa
COMPRESSIVE STRAIN, %
0/90 LAMINATE
IITRI
16-ply laminate
4-ply sandwich skin
COMPRESSIVE STRESS, MPa
COMPRESSIVE STRAIN, %

AVERAGE COMPOSITE COMPRESSION STRENGTHS :
MINI-SANDWICH TEST SPECIMEN VS. ALL-COMPOSITE TEST COUPON

COMPOSITE SYSTEM	MINI-SANDWICH SPECIMEN				COMPOSITE COUPON	
	Stacking sequence	Skin Vf (%)	IITRI* (MPa)	FPF** (MPa)	Stacking sequence	IITRI* (MPa)
S-Glass/1034	All 0°	60	2280	1430	All 0°	1400
AS4/3501-6	All 0°	65	2070	1780	All 0°	1280
AS4/APC-2	All 0°	57	1570	1630	All 0°	1100
IM8/3501-6	All 0°	65	2270	-	All 0°	1380
AS4/3501-6	0/90/90/0	62	1100	-	(0/90)4S	810

* Direct axial compression with the IITRI fixture
** Four-point flex test





SUMMARY AND CONCLUSIONS

- Conventional compression tests often underestimate composite compression strengths due to premature specimen failure.
- The mini-sandwich beam is a novel adaptation of the conventional honeycomb sandwich beam, without the disadvantages of the latter.
- Carbon fiber composite mini-sandwich specimens fail predominantly in the gage section, with failure initiating via fiber shear.
- Mini-sandwich test specimens provide more reliable and realistic estimates of composite compressive strength than all-composite coupons, especially for high compressive strength fibers.
- Unidirectional composites of AS4, IM8 and S-Glass fibers as well as AS4/epoxy cross-ply laminates fail at substantially higher stresses and strains in axial compression with the mini-sandwich specimen compared to all-composite coupons.

ACKNOWLEDGMENTS

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STRESS ANALYSIS OF EMBEDDED FIBERS

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Abstract

Recently, there has been much interest in the mechanism of fiber pull–out and its associated failure modes. This is due to the emergence of such materials as ceramic composites where the fibers provide increased toughness rather than increased strength as in organic matrix composites. Much of the work in this area is based on a paper by Aveston, Cooper, and Kelly [1]. The paper theorizes that when a brittle matrix composite is loaded, multiple fractures will occur perpendicular to the applied load at evenly spaced intervals. This spacing is dependent upon the maximum shearing stress which the fiber/matrix interface can support. Much of the current work in this field is experimental, but some analytical models for fiber pull–out do exist. The models attempt to predict the diffusion of load from the fiber to the matrix. This problem is not new and has been worked on by several leading researchers over the years. The problem itself is equivalent to the classical problem of the diffusion of load from a tension bar embedded in a three dimensional elastic medium. The problem presents formidable analytical hardship. Various approaches to the problem have been undertaken which involve different types and levels of approximations such as: modeling the bar as rigid, shearlag analysis, and energy methods. The only attempt to solve such a problem without any approximation appears to be the efforts of Pickett and Johnson [2]. However, their report considered a broken fiber embedded in a matrix and is mathematically complex and difficult to reproduce.

The objective of this work is to present a three dimensional elasticity solution to the fiber pull–out problem. The motivation behind this effort is to establish a baseline on which to compare the simplified models and to further understand the limitations of such models. In this paper, both the fiber and matrix are considered to be linearly isotropic materials and are assumed to be perfectly bonded at the interface. Because of the high temperatures often experienced by this class of composites during processing, residual thermal stresses can be significant and are incorporated into the model by assuming a linear temperature / strain relationship.

As shown in figure 1, the idealized problem consists of an isotropic fiber represented by the rod–like core of material surrounded by an isotropic matrix represented by a disk of material with inner radius, r , and an outer radius, Ω. The interface between these two constituent materials is assumed to be perfectly bonded. The radii of the fiber and matrix can be adjusted to achieve the same volume fraction as a composite laminate. In this case, we can consider the system as a representative volume element of a unidirectional composite. An axial stress of magnitude '$f / \pi \rho^2$ ' is applied to both ends of the fiber. The fourteen associated boundary conditions (listed in Figure 1) for this problem can be broken down into three groups: surface conditions, interface continuity conditions, and symmetry conditions. In order to distinguish between the fiber and matrix components, we will designate the fiber by superscript f, and the matrix material by superscript, m. The boundary value problem is axisymmetric. In this case, the stress components are independent of the angle θ and it follows that all derivatives with respect to θ vanish. The shearing stresses τ_{rz} and $\tau_{r\theta}$ also vanish due to symmetry. Therefore, the equilibrium equations [3] reduce to:

$$\frac{\partial \sigma_r}{\partial r} + \frac{\partial \tau_{rz}}{\partial z} + \frac{\sigma_r - \sigma_\theta}{r} = 0 \tag{1}$$

$$\frac{\partial \tau_{rz}}{\partial r} + \frac{\partial \sigma_z}{\partial z} + \frac{\tau_{rz}}{r} = 0 \tag{2}$$

The engineering strain components, for axially symmetrical deformation, are:

$$\varepsilon_r = \frac{\partial u_r}{\partial r} \qquad \varepsilon_\theta = \frac{u_r}{r} \qquad \varepsilon_z = \frac{\partial u_z}{\partial z} \qquad \gamma_{rz} = \frac{\partial u_r}{\partial z} + \frac{\partial u_z}{\partial r} \tag{3}$$

Using the notation that 1,2, 3, and 6 correspond to z, r, θ, and rz, respectively, Hooke's Law is:

$$\sigma_i = C_{ij}(\varepsilon_j - \varepsilon_j^0) \qquad i, j = 1, 2, 3, 6 \tag{4}$$

where,

$$\varepsilon_j^o = \alpha_j \, \Delta T = \text{thermal strain}$$

Equations (1)–(4) represent the complete system of governing field equations. Because of the isotropic nature of the constituents, we can simplify the problem using Love's stress function [4] and reduce the field equations for each material to a single partial differential equation:

$$\nabla^4 \varnothing = 0 \tag{5}$$

where,

$$\nabla^2 = \frac{\partial^2}{\partial z^2} + \frac{\partial^2}{\partial r^2} + \frac{\partial}{r\,\partial r}$$

The stresses and displacements can then be derived from:

$$\sigma_r = \frac{\partial}{\partial z}[\nu\, \nabla^2 \varnothing - \frac{\partial^2 \varnothing}{\partial r^2}] \tag{6}$$

$$\sigma_\theta = \frac{\partial}{\partial z}[\nu\, \nabla^2 \varnothing - \frac{\partial \varnothing}{r\,\partial r}] \tag{7}$$

$$\sigma_z = \frac{\partial}{\partial z}[(2 - \nu)\, \nabla^2 \varnothing - \frac{\partial^2 \varnothing}{\partial z^2}] \tag{8}$$

$$\tau_{rz} = \frac{\partial}{\partial r}[(1 - \nu)\, \nabla^2 \varnothing - \frac{\partial^2 \varnothing}{\partial z^2}] \tag{9}$$

$$2G\, U_r = - \frac{\partial^2 \varnothing}{\partial r\, \partial z} \tag{10}$$

$$2G\, U_z = 2(1 - \nu)\nabla^2 \varnothing - \frac{\partial^2 \varnothing}{\partial^2 z} \tag{11}$$

It should be emphasized that Equation 5 must be solved independently for each constituent material. The appropriate stress functions are:

$$\varnothing^f = \sum_{i=1}^{\infty} [\, A_{1i}\, I_0(k_i r) + A_{2i} k_i r\, I_1(k_i r)]\, \sin(k_i z)$$
$$+ \sum_{j=1}^{\infty} [C_{1j} z \cosh(\alpha_j z) + C_{2j} \sinh(\alpha_j z)] e^{-\alpha_j L}\, J_0(\alpha_j r)$$
$$+ F_1 r^2 z + F_3 z^3 \tag{12}$$

$$\varnothing^m = \sum_{i=1}^{\infty} [\, B_{1i}\, K_0(k_i r) + B_{2i} k_i r\, K_1(k_i r) + B_{3i}\, I_0(k_i r) + B_{4i} k_i r\, I_1(k_i r)]\, \sin(k_i z)$$
$$+ \sum_{j=1}^{\infty} [D_{1j} z \cosh(\alpha_j z) + D_{2j} \sinh(\alpha_j z)] e^{-\lambda_j L} [J_0(\lambda_j r) + \mu_j Y_0(\lambda_j r)]$$
$$+ G_1 r^2 z + G_2 z \ln(r) + G_3 z^3 \tag{13}$$

where J and Y are Bessel functions of first and second kind, respectively. I and K are the corresponding modified Hyperbolic Bessel functions. Both stress functions are suitably bounded on the intervals considered and can be shown to be solutions to the biharmonic equation by substitution into Equation 5. It is interesting to note that the basic problem is solved by the superposition of three well known solutions. The first summation term in the stress function is the solution to an infinite fiber embedded in an infinite matrix. The second summation term is similar to the solutions of a penny shaped crack and the application of pressure to free surfaces of elastic plates. The last term of the stress function is the far field or steady state solution. For numerical convenience, the coefficient of each term in the solution containing these functions is multiplied by a negative exponential power as shown in Equations 12 and 13. The constants k_i and α_j are eigenvalues defined by the following relationships:

$\sin(k_i l) = 0$	$i=1, 2, 3, \ldots$	(14)
$J_1(\alpha_j \rho) = 0$	$j=1, 2, 3, \ldots$	(15)

making each set of functions orthogonal over the intervals 0 to L and 0 to ρ, respectively. The constants μ_j and λ_j are defined by the relationships:

$\mu_j = -J_1(\lambda_j \rho) / Y_1(\lambda_j \rho)$	$j=1, 2, 3, \ldots$	(16)
$J_1(\lambda_j \Omega) + \mu_j\, Y_1(\lambda_j \Omega) = 0$	$j=1, 2, 3, \ldots$	(17)

which defines a complete set of orthogonal functions on the interval of ρ to Ω. Upon substitution of Equations 12 and 13 into Equations 6–11, the boundary conditions can be satisfied by the resultant expressions with judicious use of the orthogonality of the sine, cosine, and Bessel functions. The system of equations is then truncated and solved with a computer using standard library routines.

Figure 2 shows the typical results for the hoop stress as function of z for constant values of r. The results are smooth at 1.10 ρ, but become increasingly oscillatory as the interface is approached. This is a result of a Gibbs Phenomenon as discussed in the third viewgraph . However, the radial stress is not oscillatory and is used to perform a convergence study as shown in Figure 3. Figures 4–7 show the results for a SCS6 fiber in a 7740 glass matrix after a refinement process to remove the Gibbs Phenomenon.

REFERENCES

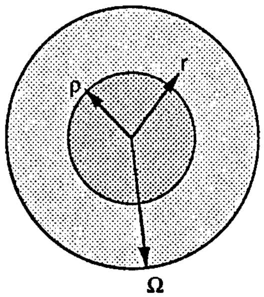
1. Aveston, J., Cooper, G.A., and Kelly, A. " Single and Multiple Fracture" Conf. Proc. Nat. Phys. Lab., IPC Science and Technology Press Ltd, 4 Nov 1971.

2. Pickett, G. and Johnson, M.W. "Analytical Procedures for Predicting the Mechanical Properties of Fiber Reinforced Composites", Technical Report AFML–TR–5–220, 1 Oct 1967.

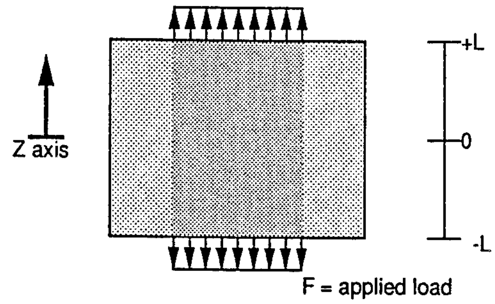
3. Timoshenko,, S. and Goodier, J.N. *Theory of Elasticity*, McGraw–Hill Book Company, Inc., New York, pp. 179, 1951.

4. Love, A.E.H. *A Treatise on the Mathematical Theory of Elasticity*, 4th ed., Dover, New York, 1944.

FIGURE 1- SYSTEM DIAGRAM



FIBER-MATRIX TOP VIEW



FIBER-MATRIX SIDE VIEW

BOUNDARY CONDITIONS

interface continuity

$U_z^m(\rho, z) = U_z^f(\rho, z)$	$0 \le z \le L$	(1.b)
$U_r^m(\rho, z) = U_r^f(\rho, z)$	$0 \le z \le L$	(2.b)
$\sigma_r^m(\rho, z) = \sigma_r^f(\rho, z)$	$0 \le z \le L$	(3.b)
$\tau_{rz}^m(\rho, z) = \tau_{rz}^f(\rho, z)$	$0 \le z \le L$	(4.b)

free surface

$\sigma_r^m(\Omega,z) = 0$	$0 \le z \le L$	(5.b)
$\tau_{rz}^m(\Omega,z) = 0$	$0 \le z \le L$	(6.b)
$\tau_{rz}^f(r,L) = 0$	$0 \le r \le \rho$	(7.b)
$\sigma_z^f(r,L) = f / \pi\, \rho^2$	$0 \le r \le \rho$	(8.b)
$\tau_{rz}^m(r,L) = 0$	$\rho \le r \le \Omega$	(9.b)
$\sigma_z^m(r,L) = 0$	$\rho \le r \le \Omega$	(10.b)

symmetry

$U_z^f(r, 0) = 0$	$0 \le r \le \rho$	(11.b)
$\tau_{rz}^f(r, 0) = 0$	$0 \le r \le \rho$	(12.b)
$U_z^m(r, 0) = 0$	$\rho \le r \le \Omega$	(13.b)
$\tau_{rz}^m(r, 0) = 0$	$\rho \le r \le \Omega$	(14.b)

FIGURE 2- HOOP STRESS FOR ΔT = -.5555 °C (UNREFINED)

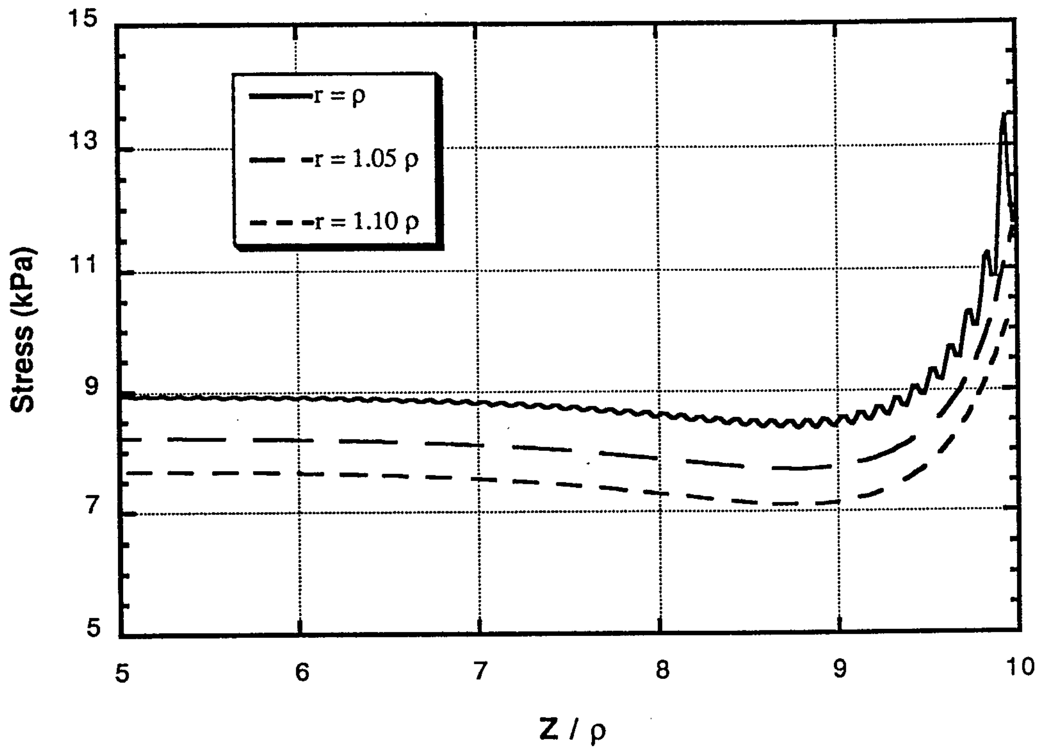
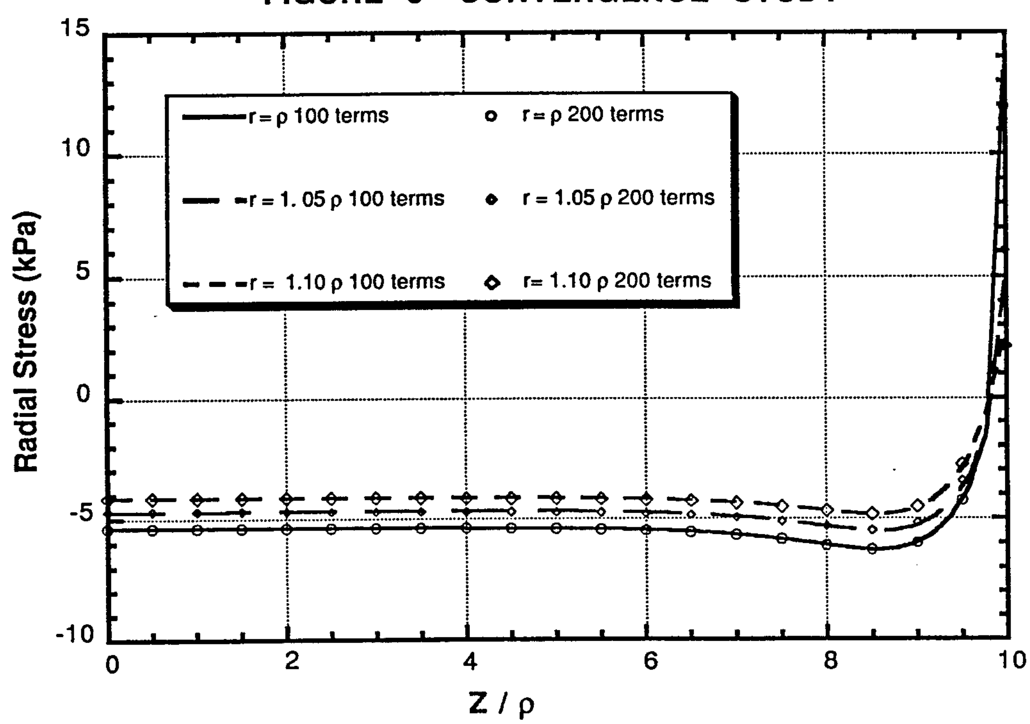


FIGURE 3- CONVERGENCE STUDY



Gibbs Phenomenon

- Period of oscillation (= L / # terms, symmetric case)

- Amplitude of the oscillation gets larger as the free surface is approached (approx. 9%)

- Does not appear to be an oscillatory singularity

  -- Radial stress is non-oscillatory

  -- Eigenfunction expansion type solutions do not predict oscillatory behavior

- 3 ways to handle

  -- Take more terms

  -- "Patch" the solution with existing eigenfunction solutions

  -- Use only those point which are known to be *correct*

Material Properties / Goemetry

- 7740 glass matrix reinforced with a Silicon Carbide (SCS6) fiber

  — $E_f = 421$ GPa $\quad \nu_f = 0.2 \quad \alpha_f = 3.25$ m / ° C

  — $E_m = 63$ GPa $\quad \nu_m = 0.2 \quad \alpha_m = 3.50$ m / ° C

- length, fiber radius, and outer radius of the matrix

  —$L = 10\rho \quad \rho = 0.1 \quad \Omega = 0.2\rho$

FIGURE 4- RADIAL STRESS FOR ΔT = -.5555 °C

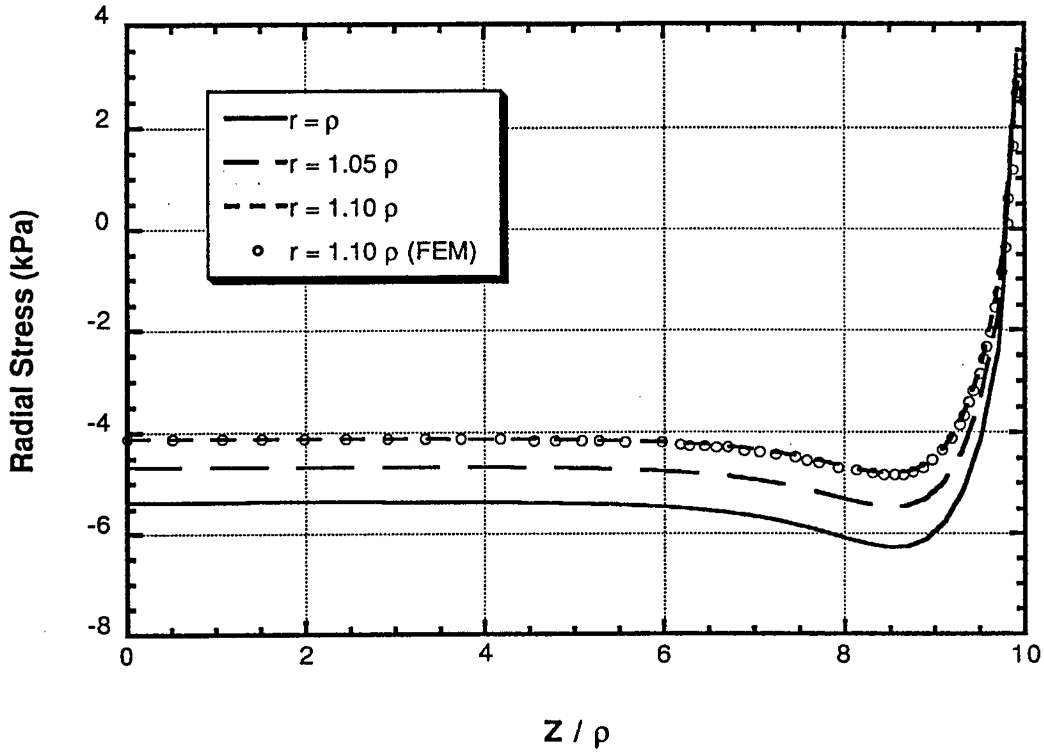


FIGURE 5- HOOP STRESS FOR ΔT = -.5555 °C

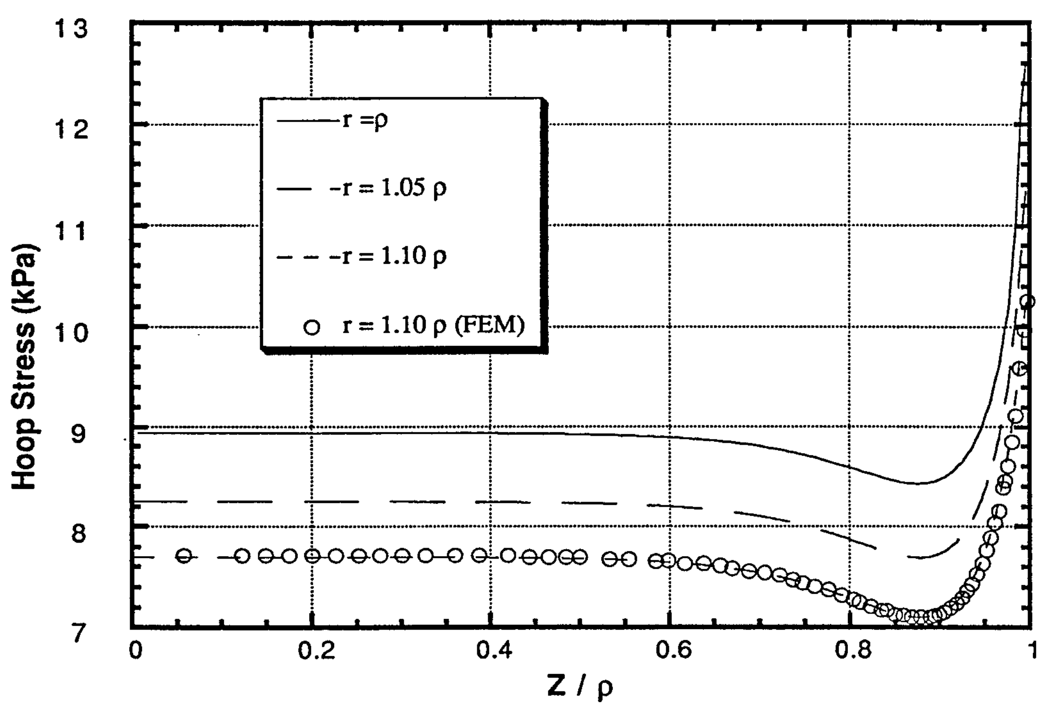


FIGURE 6- AXIAL STRESS FOR Δ T = -.5555 °C

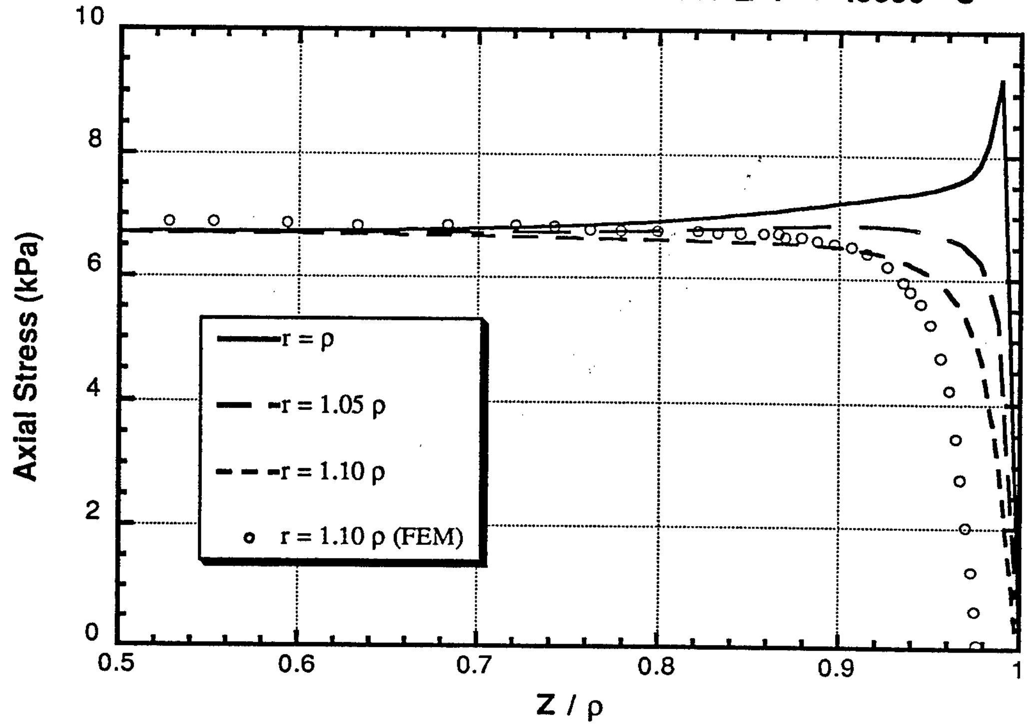
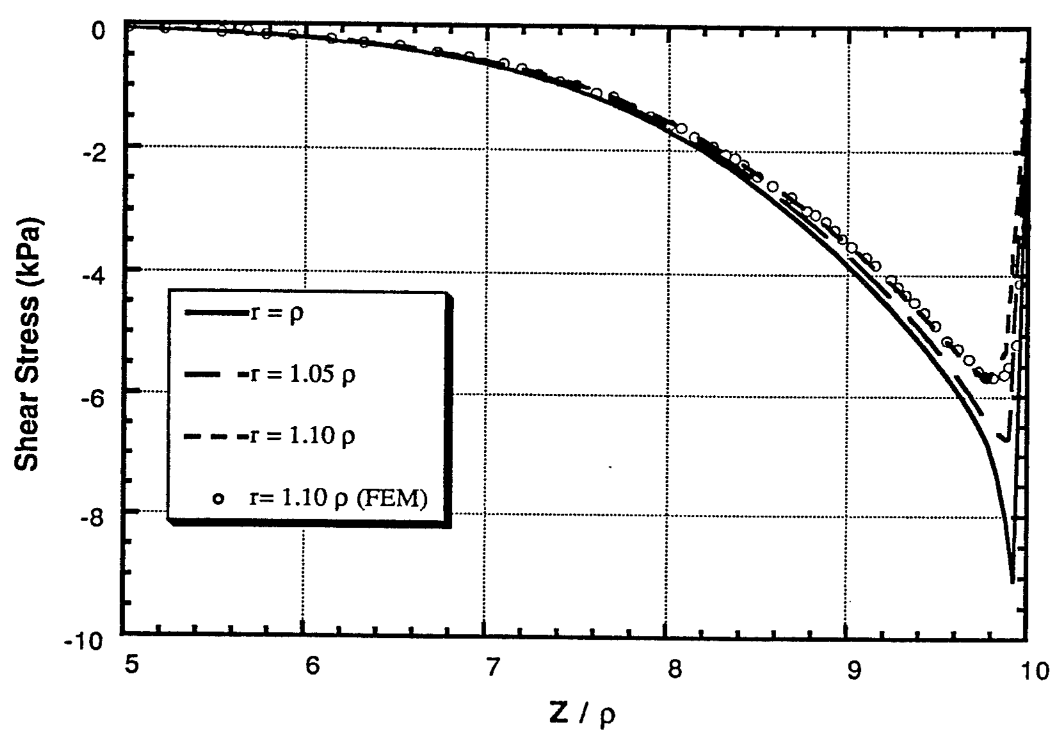


FIGURE 7- SHEAR STRESS FOR ΔT = -.5555 °C



Conclusions

- An analytical solution for the fiber/pullout type phenomenon has Been presented

    -- Mathematically correct, but their is no basis to compare to

    -- Captures the singular nature of the stresses

    -- Correlates with the results of finite elements

- Drawbacks

    -- Gibb's Phenomenon

    -- A system of equations with an infinite number of equations and an infinite number of unknowns

    -- Mathematically complex

- The bottom line

<u>Much has been learned about the nature of the problem, but much more needs to be done.</u>

HIGH TEMPERATURE FATIGUE OF A TITANIUM ALUMINIDE COMPOSITE

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Titanium aluminide composites are currently the prime candidate materials for application such as hypervelocity aircraft and advanced turbine engines. Prior to using these composites effectively in high performance structures, it is essential to develop a complete understanding of their fatigue behavior under realistic operating conditions. The material tested in this study is the SCS-6/Ti-24Al-11Nb composite, which is an alpha-two titanium aluminide alloy, unidirectionally reinforced with 35% volume fraction silicon carbide fibers. Past research of this composite is limited to a fatigue and fracture study conducted by the Allison Gas Turbine Division of General Motors. This study focuses on high cycle fatigue of this composite at 815°C. The objective of this study is to evaluate the composite behavior under these conditions by considering the effects of frequency, mean stress and stress range on fatigue life and damage evolution.

Straight-sided specimens were machined from 8-ply composite panels using an abrasive water jet technique. These panels were fabricated by the foil-fiber-foil method which utilized molybdenum crossweave for improved fiber alignment. Specimens were tested in a fully automated high cycle fatigue test system. Specimens were heated by quartz lamps and temperature was monitored from a control sample held adjacent to the test specimen.

Several techniques were used during testing to detect damage. Physical observations were made by taking surface replicas periodically during testing of selected tests and by fracture surface analyses. Electric potential difference was monitored as an indication of matrix damage. Elastic modulus was monitored during testing as an indication fiber damage. The residual strength of some specimens was measured as an overall indication of composite damage after the specimens had undergone fatigue cycling.

Tests were conducted at a constant maximum cyclic stress of 430 MPa, the applied loading frequency and stress ratio were varied to study the effects of frequency, mean stress and stress range. Primary tests were conducted at frequencies of 30 Hz and 300 Hz with stress ratios of 0.1, 0.3, 0.5 and 0.8. One test was conducted at 1 Hz with a stress ratio of 0.1 and other tests were conducted at 0.01 Hz with stress ratios of 0.1, 0.5 and 0.8.

The fatigue tests vary over about six orders of magnitude in cycles-to-failure. At each stress ratio or stress range level the fatigue data are ordered according to frequency. At a given stress ratio, the number of cycles-to-failure is approximately proportional to frequency. The fatigue data at each stress ratio coalesce when plotted on a time-to-failure basis. The time-to-failure (failure life) for all the tests varies over about two orders of magnitude, compared with six orders of magnitude exhibited in cycles-to-failure. On the time scale the ordering of the fatigue data at each stress ratio is random with respect to frequency. The results indicate that the failure life of the composite is time dependent at each stress ratio and loading frequency has no significant effect on failure life. However, failure lives vary with stress ratio due to both mean stress and stress range effects on fatigue.

Surface replicas were taken to determine the evolution of damage on the specimen surface for three stress ratio conditions. Cracks initiated along the specimen edge were determined to be directly related to the molybdenum crossweave used in fabrication of the composite. Surface damage evolution is dependent on the stress condition. Damage evolution is attributed to both environmental degradation and mechanical loading. Fracture surface analyses indicate that at low stress ratios self-similar type crack growth occurs through the matrix. At high stress ratios fatigue damage propagates on several horizontal planes within the thickness of the specimen, resulting in a stepped-type fracture surface.

Electric potential difference measurements reflect the accumulation of damage in the matrix; however, this determination was only possible by correlating the trends with the surface replication results. Elastic modulus exhibits only a slight reduction over life for tests at each stress ratio. The residual strength results indicate that significant losses in composite strength occur while the elastic modulus remains relatively constant. This indicates that significant fiber damage occurs which cannot be measured by elastic modulus.

A phenomenological model was developed assuming a linear life fraction summation, based on the observations of damage evolution at the various stress conditions. The model predicts the failure life of the composite fairly well, accounting for perturbations in failure life at the various stress ratios due to mean stress and stress range effects.

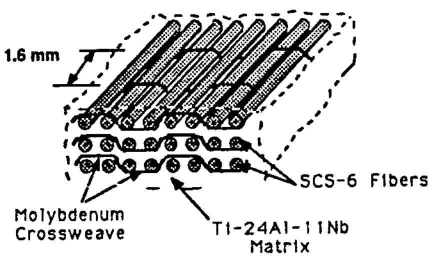
Material

SCS-6/Ti-24Al-11Nb

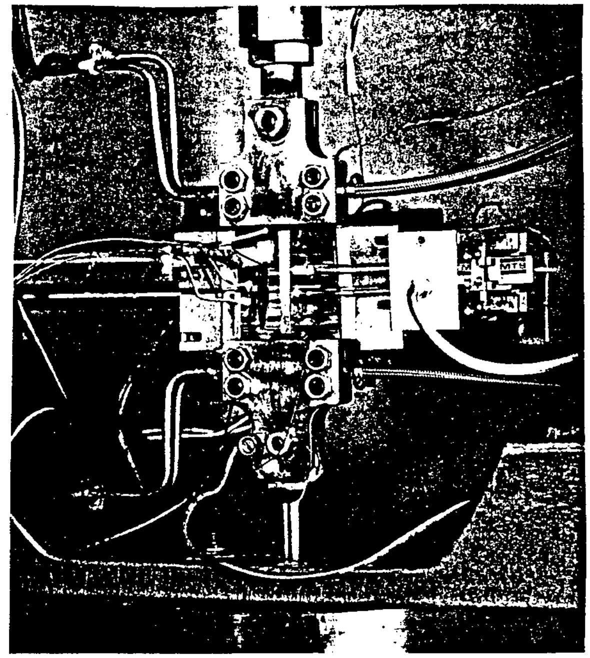
Unidirectional

8 - Ply

Volume Fraction of Fibers ≈ 35 %



Test Setup



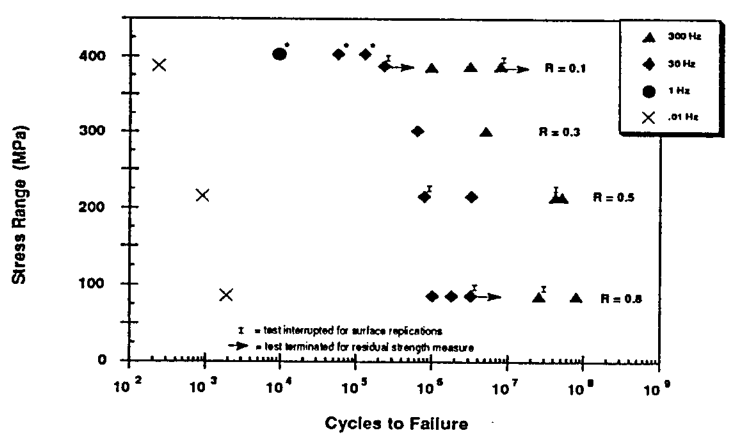
Physical Observations

- Surface Replication
- Fractography

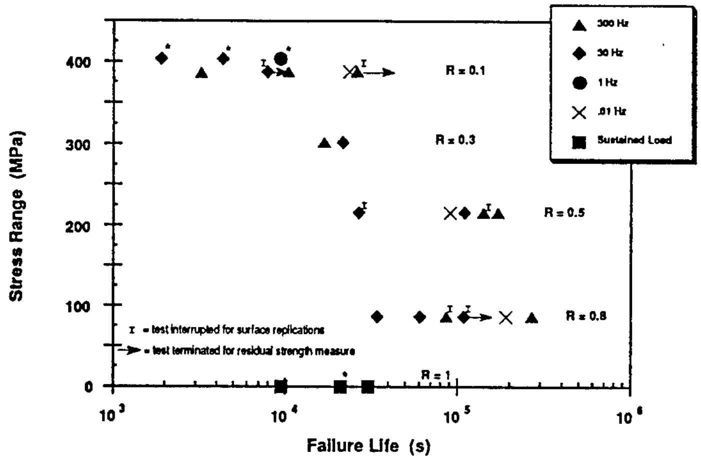
Damage Monitoring

- Electric Potential Difference
- Elastic Modulus
- Residual Strength

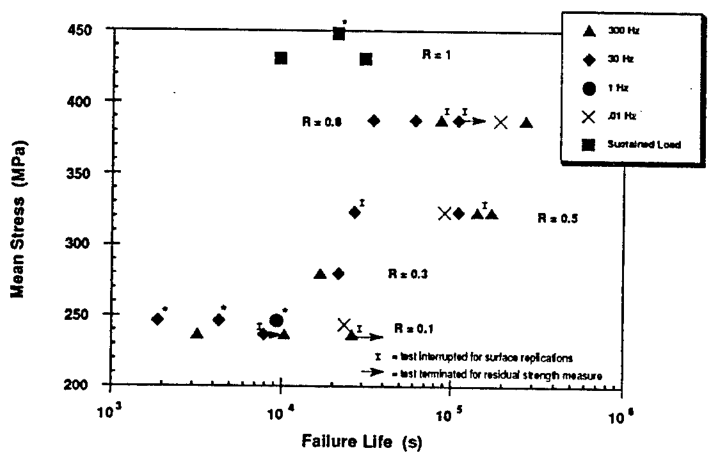
SCS-6/Ti-24Al-11Nb
815°C, Max Stress = 430 MPa (* 450 MPa)



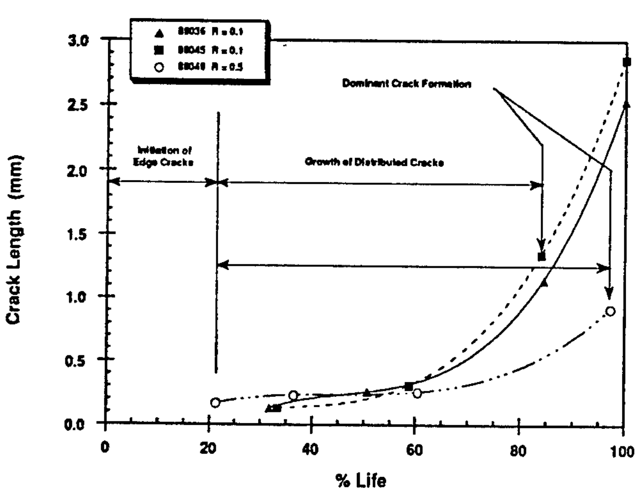
SCS-6/Ti-24Al-11Nb
815°C, Max Stress = 430 MPa (* 450 MPa)



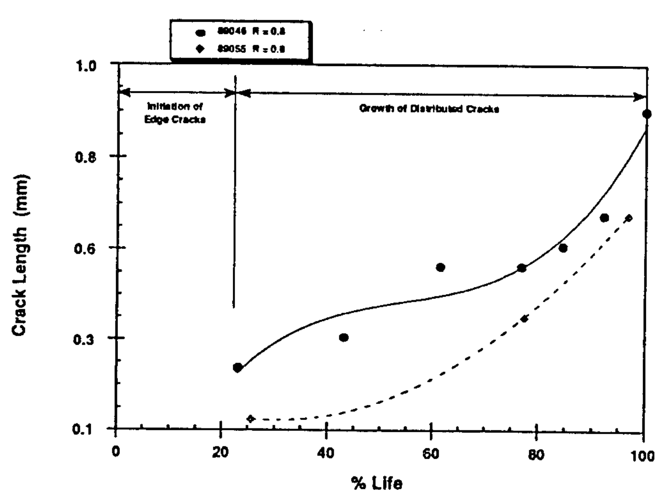
SCS-6/Ti-24Al-11Nb
815°C, Max Stress = 430 MPa (* 450 MPa)



Surface Replication Results



Surface Replication Results



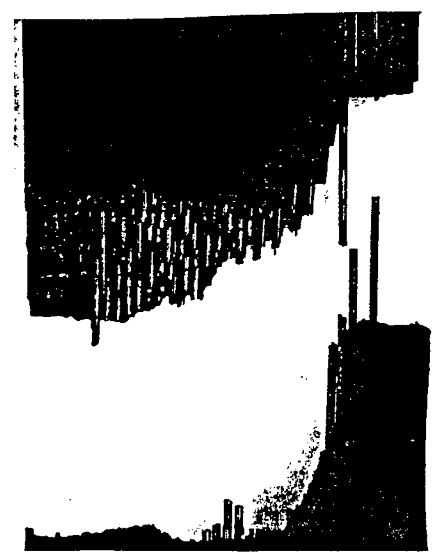
Failure Mode as a Function of Stress Ratio (R)

R = 0.1
σ_{mean} = 273 MPa

R = 0.3
σ_{mean} = 297.5 MPa

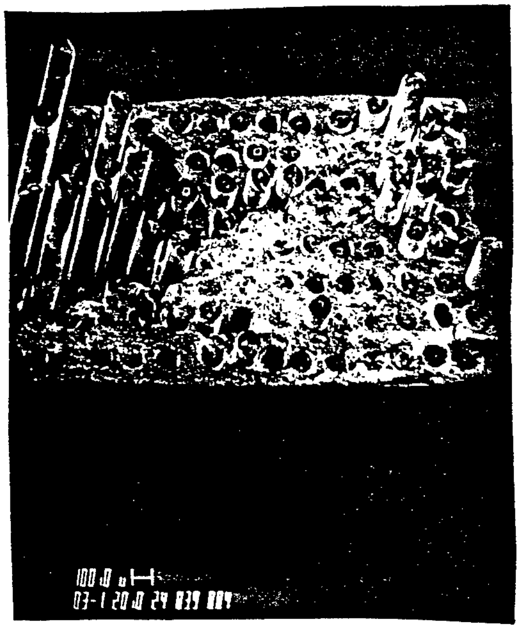
Failure Mode as a Function of Stress Ratio (R)

R = 0.8
σ_{mean} = 388 MPa

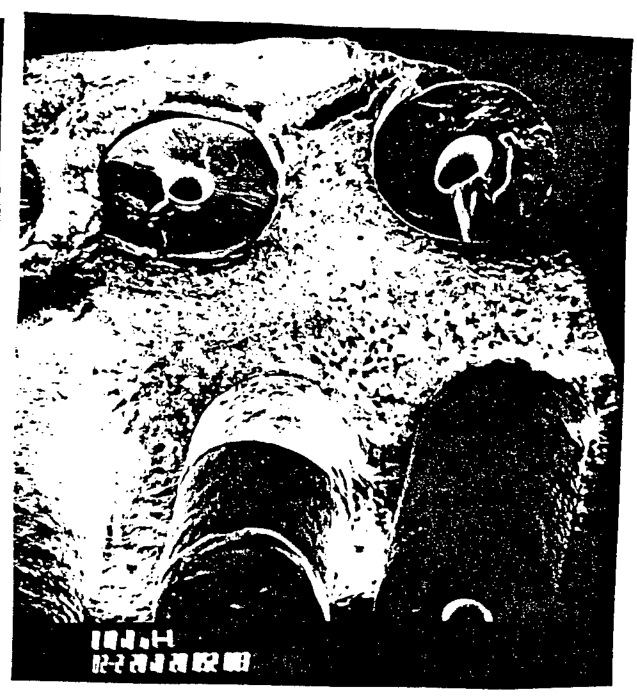


R = 1
σ_{mean} = 430 MPa

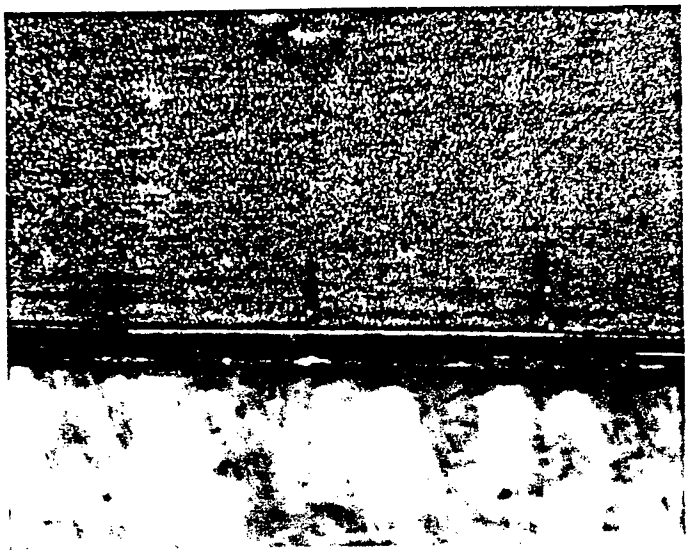
Typical Fracture Surface for Low Stress Ratio (R = 0.1,0.3,0.5)



Typical Fracture Surface for High Stress Ratio (R = 0.8,1)

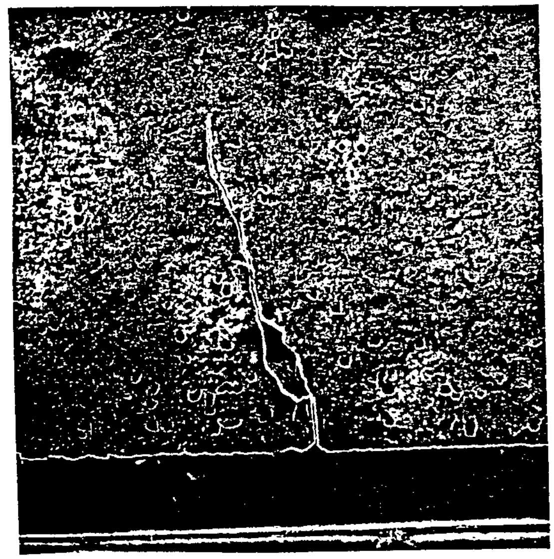


Periodic Cracks on Specimen Surface



1 mm

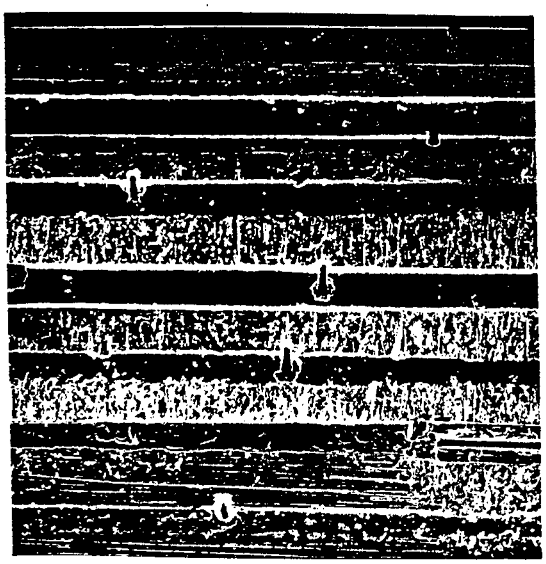
Surface View - Low Mag



100 μm

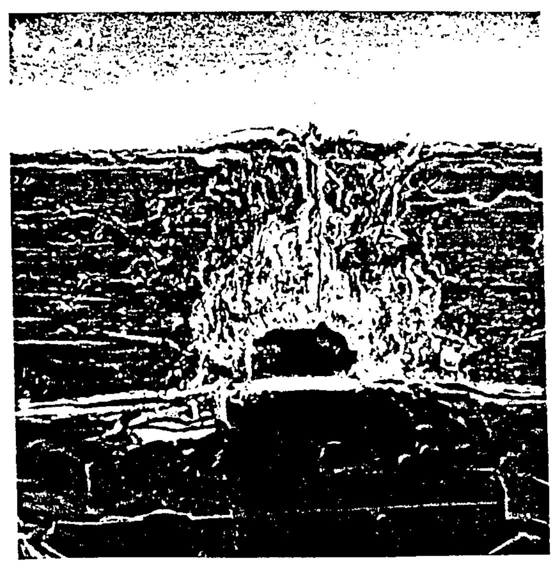
Surface View - High Mag

Cracks Initiated From Molybdenum Cross-Weave



400 μm

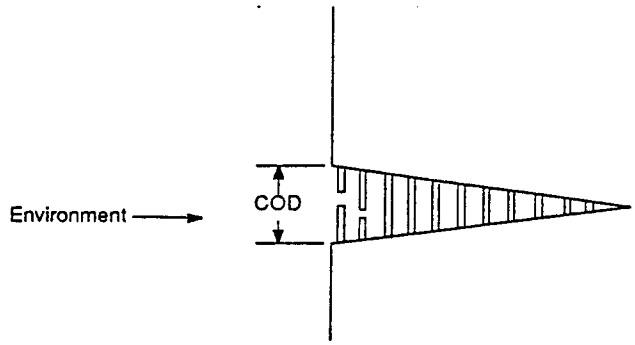
Edge View - Low Mag



50 μm

Edge View - High Mag

Damage Evolution

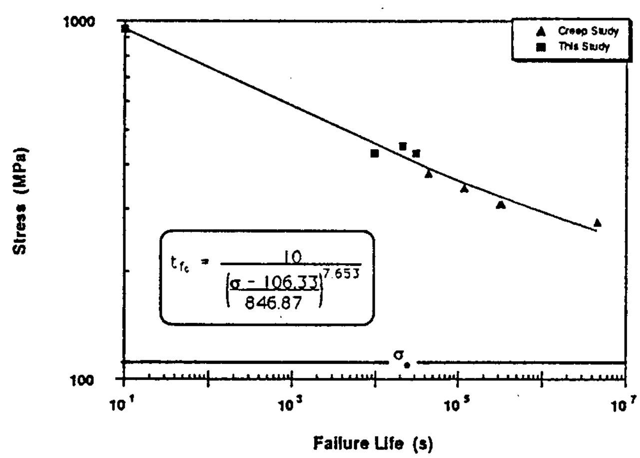


Damage evolution = f(time, ΔCOD)

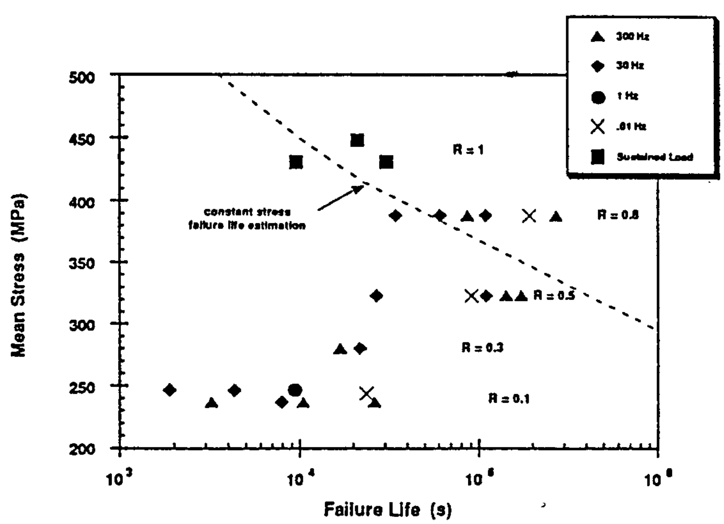
High $\Delta\sigma$: large ΔCOD
matrix crack propagates through thickness
fibers through the thickness are damaged

Low $\Delta\sigma$: small ΔCOD
matrix cracks propagate through individual plies
fibers through the thickness are damaged on several horizontal planes

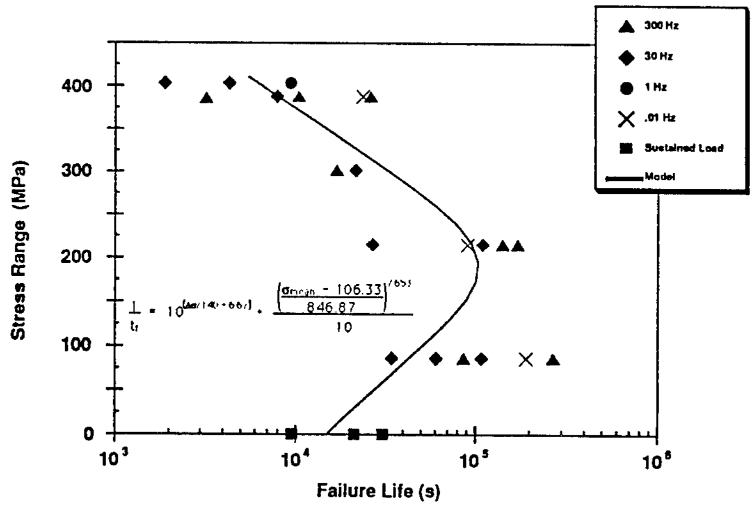
Sustained Load Modeling



Sustained Load Modeling



Life-Fraction Modeling



Conclusions

- Damage initiates at the molybdenum crossweave sites.
- Fatigue Life on a cycles-to-failure basis exhibits a strong frequency dependence.
- Frequence dependence disappears when the data are displayed on a time-to-failure basis, indicating that damage evolution is time-dependent.
- Stress range controls failure life behavior for low stress ratio conditions.
- Mean stress controls the failure life behavior for high stress ratio conditions.
- Damage evolution is dependent on stress ratio condition:
  self-similar crack growth at low stress ratios,
  stepped crack growth at high stress ratios.
- Fiber strength is degraded by exposure to the detrimental environment during fatigue loading. Fiber modulus remains relatively constant throughout life.
- A phenomenological failure life model was developed which accounts for the competing stress range and mean stress effects.

THERMAL MECHANICAL FATIGUE ANALYSIS OF TITANIUM ALUMINIDE MMC

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Metals and Ceramics Division

ABSTRACT

The mismatch of material properties of the fiber and matrix in a MMC generates residual stresses during processing and thermal loading. The determination of the mechanical stress and strain ranges developed within the matrix during thermal and mechanical loading is required to understand the material response. As one considers thermal-mechanical modeling of MMC, various complexities develop in the analysis. For example, a simple elastic analysis of the MMC consolidation processing can predict matrix stresses that exceed the material yield stress. To model the material yielding behavior more realistically requires a plasticity analysis. In addition, most materials exhibit a temperature dependent material behavior. To account for this material behavior, additional terms are added to the isothermal plasticity theory. To handle the increasing complexity of thermal and mechanical behavior of MMC, the finite element method is well suited to satisfy the model's loading and boundary conditions with the appropriate plasticity theory.

The thermal plasticity theory with temperature-dependent material properties has been implemented into the finite element package MAGNA [1] and is used to analyze MMC models subjected to thermal and mechanical loads. This presentation includes many details of the classical plasticity theory [2-4] and finite element model to illustrate the many aspects and complexities encountered during the TMF modeling of MMC. The resulting stress and strain predictions are then used to show some of the limitations of simpler models.

This analysis considers a titanium aluminide matrix reinforced with unidirectional silicon carbide fibers. The titanium aluminide of interest does exhibits temperature dependent behavior as the elastic modulus, yield stress, plastic hardening rate, and the coefficient of thermal expansion are all temperature dependent. The plasticity approach should account for these temperature effects to realistically model material behavior. This analysis considers a classical elastic-plastic theory with the addition of temperature dependent material properties. The plasticity model consists of the Von-Mises yield criteria, the consistency condition, and the associated flow rule. The basic equations which illustrate the general approach to the thermal plasticity theory are presented here.

To model a unidirectional MMC, two different finite element models are considered: a square cell and a concentric cylinder model. The square cell model is composed of three dimensional solid elements to represent one quarter of a matrix/fiber geometry in a uniform array of fibers. The concentric cylinder model is composed of two dimensional axisymmetric finite elements to represent one unit of a "bundle" of concentric cylinders. A general plane strain (uniform strain) condition is imposed along the axial direction, with contact elements at the fiber/matrix interface. The contact model allows separation of matrix from fiber, but does not account for a finite bond or friction between the fiber and matrix [5]. Additional constraints are added to the square cell model to satisfy symmetry conditions.

The results of three different analyses of MMC's are presented. The first analysis shows a comparison of process-induced stress prediction for a square cell model between assumed thermal-elastic and elastic-plastic material behaviors. Then, two analyses are presented for concentric cylinder models subjected to I.) mechanical fatigue at a constant elevated temperature and II.) thermal fatigue with a constant axial load. These loading conditions are the same as those found in thermal mechanical experiments conducted within the materials lab [6]. These load cases are simplified cases of thermal-mechanical fatigue; however, they illustrate many aspects and levels of complexity encountered during thermal mechanical analysis of MMC.

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[3] Hunsaker B., Haisler W.E., and Stricklin J.A. "On the Use of Two Hardening rules of Plasticity in Incremental and Pseudo Force Analysis", **Constitutive Equations in Viscoplasity - Computational and Engineering Aspects**, AMD - Vol. 20, ASME 1976.

[4] Bushnell D. "A Subincremental Strategy for Solving Problems Involving Large Deflections, Plasticity and Creep," **Constitutive Equations in Viscoplasity - Computational and Engineering Aspects**, AMD - Vol. 20, ASME 1976.

[5] Nimmer R. P., Bankert R.J., Russell E. S. and Smith G. A., "Micromechanical Modeling Fiber/Matrix Interface Effects in SiC/Ti Metal Matrix Composites" Presented at AAM Materials Week Conference, October, 1989, Indianapolis Indiana.

[6] Russ S.M. and Nicholas T. "Thermal and Mechanical Fatigue of Ti-Aluminides MMC, Presented at **Titanium Aluminide Composite Workshop**, May 16-18, 1990, Orlando Florida

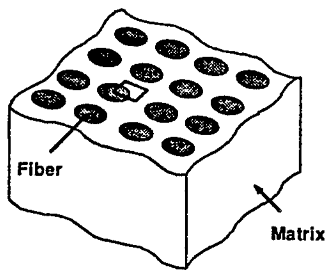
Outline

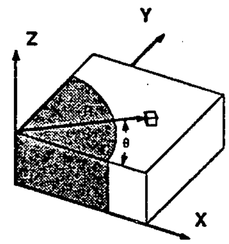
1. Objectives
2. MMC models and material properties
3. Thermal elastic-plastic theory with temperature dependent material properties
4. Elastic and elastic-plastic comparisons for residual stress state due to thermal processing
5. Mechanical cycling at elevated temperature
6. Thermal cycling with applied axial load
7. Conclusions

Objective

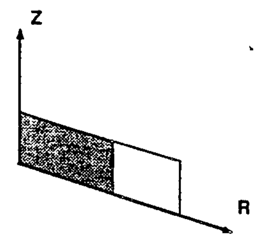
- To determine stress states in MMC due to thermal processing and subsequent thermal and/or mechanical loads.
  - To account for plasticity and temperature dependent material properties in stress analysis.
  - To implement theory into a finite element package.
  - Expand theory to include effects of time dependent constitutive models. (e.g., creep)
  - To implement unified inelastic strain theory (Bodner/Partom) into finite element package.

MMC Model Geometries



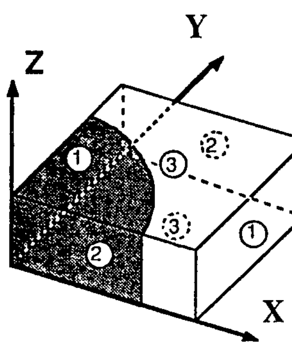


Square Cell Model



Axisymmetric Model

Boundary Conditions for Typical Cell



on 1

$$u = u_o, \quad \sigma_{xy} = \sigma_{xz} = 0, \quad \int_1 \sigma_{xx}\, dA = 0$$

on 2

$$v = v_o, \quad \sigma_{yx} = \sigma_{yz} = 0, \quad \int_2 \sigma_{yy}\, dA = 0$$

on 3

$$w = w_o, \sigma_{zx} = \sigma_{zy} = 0, \quad \int_3 \sigma_{zz}\, dA = 0$$

Thermal Plasticity Summary

Von-Mises Yield Function

$$F = \frac{3}{2}(s_{ij}-\alpha_{ij})(s_{ij}-\alpha_{ij}) - k^2 = 0$$

with

$$s_{ij} = \sigma_{ij} - \frac{\sigma_{kk}}{3}\delta_{ij}$$

and, $\alpha_{ij} = 0$, for isotropic hardening

Consistency Condition

$$\frac{\delta F}{\delta \sigma_{ij}} d\sigma_{ij} - 2k\, dk = 0$$

with

$$d\sigma_{ij} = E_{ijlm}\,(d\varepsilon^{tot}_{lm} - d\varepsilon^{th}_{lm} - d\varepsilon^{pl}_{lm}) + dE_{ijlm}(\varepsilon^{tot}_{lm} - \varepsilon^{th}_{lm} - \varepsilon^{pl}_{lm})$$

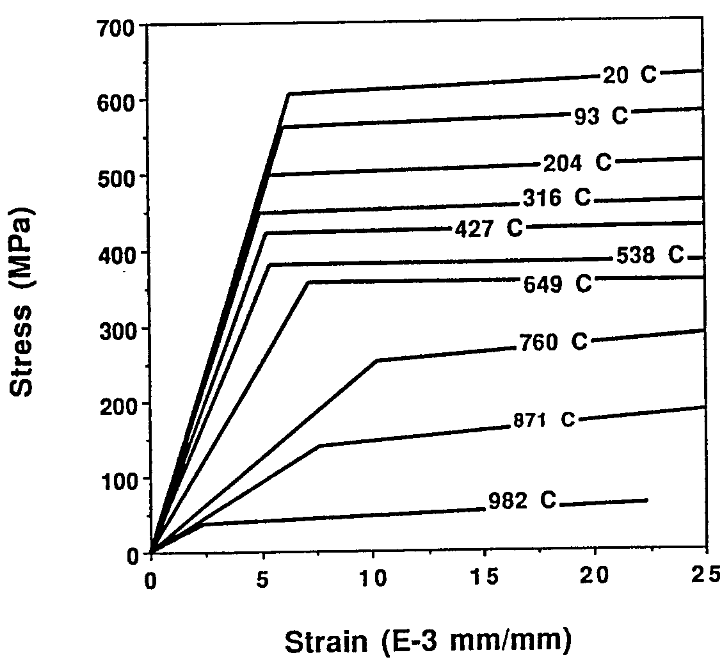
and

$$dk = dY_o + H\; d\varepsilon^{pl} + dH\,\varepsilon^{pl}$$

Plastic flow rule

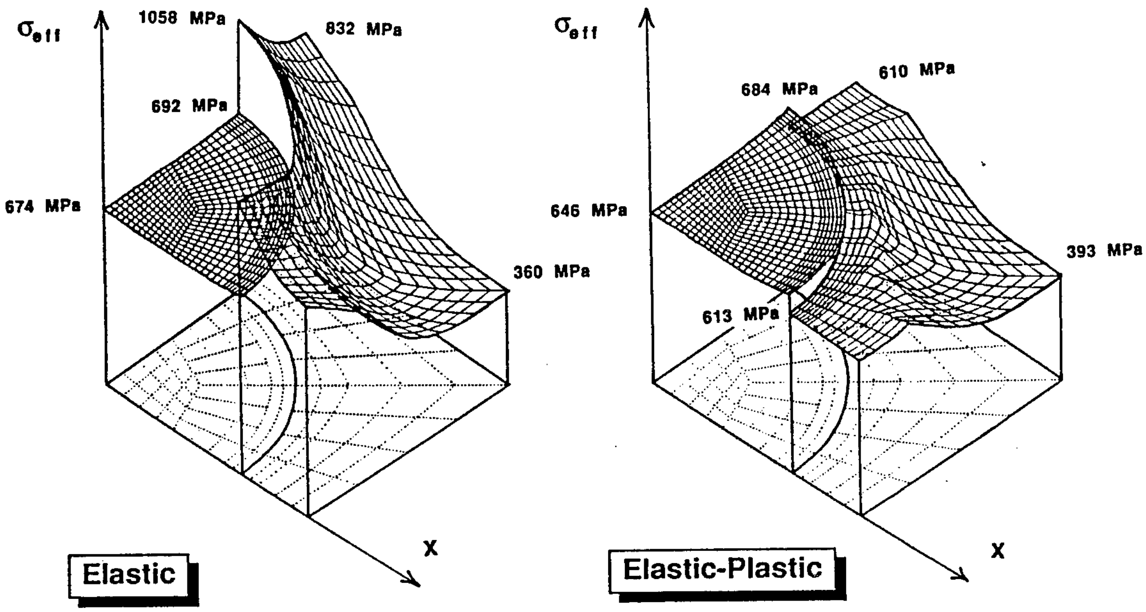
$$d\varepsilon^{pl}_{ij} = d\lambda \,\frac{\delta F}{\delta \sigma_{ij}}$$

Uniaxial Stress/Strain Data for Ti3Al

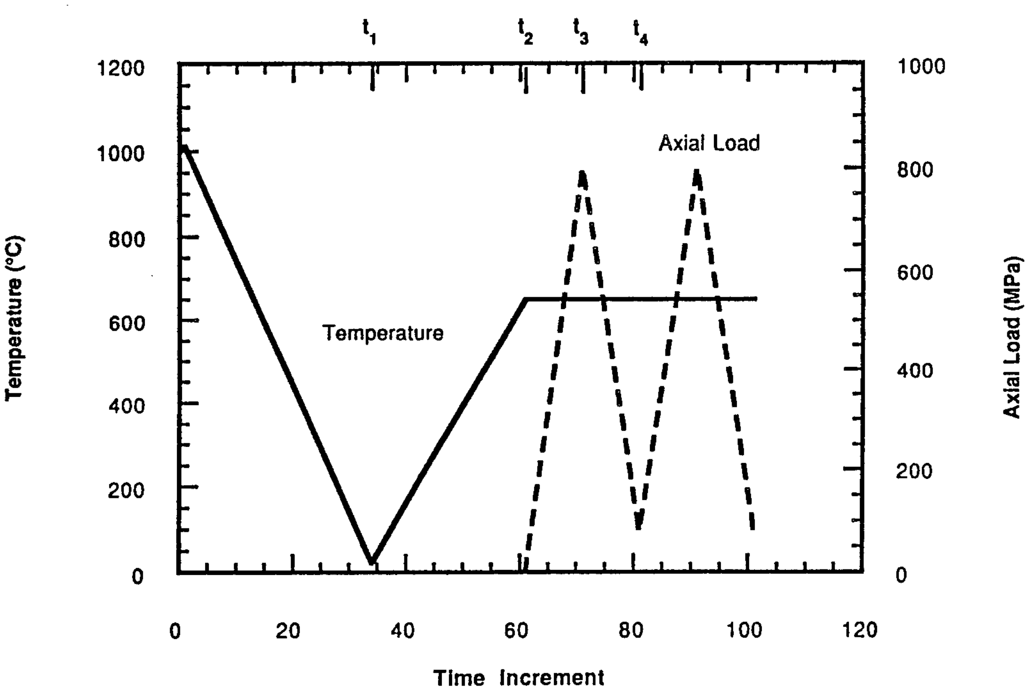


$\upsilon = 0.3$

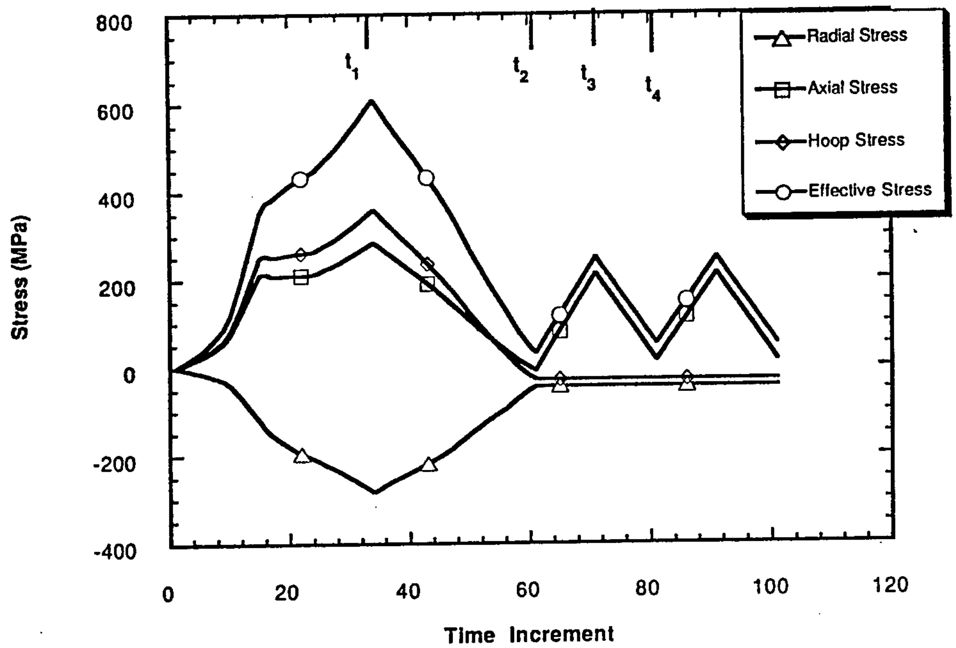
Comparison of Results from Elastic and Elastic-Plastic Analyses



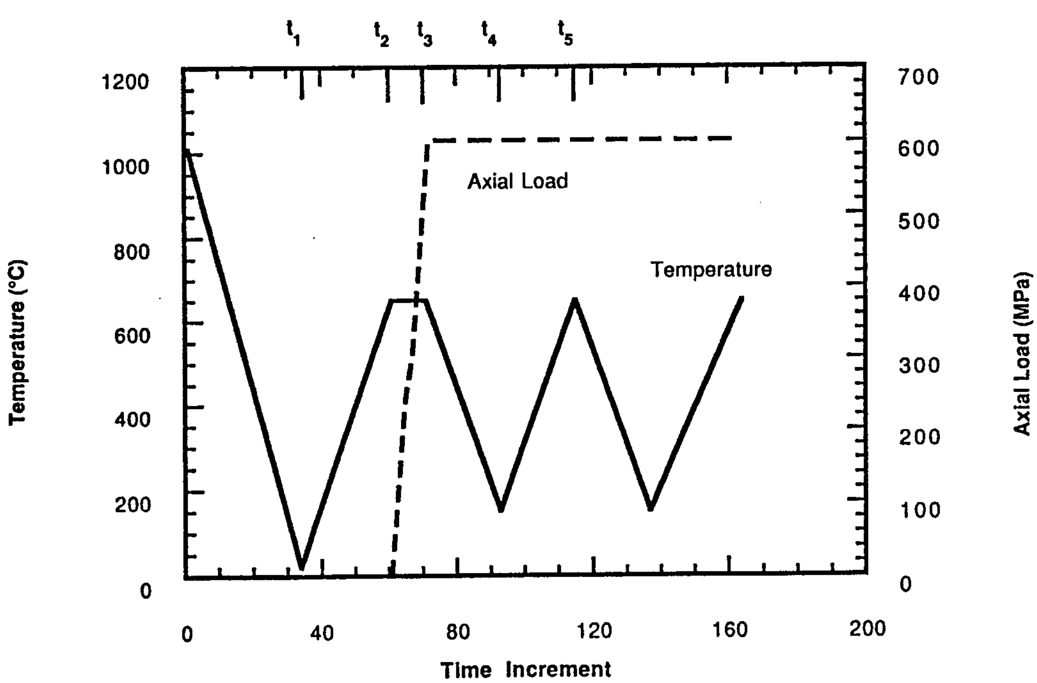
Thermal and Mechanical Loading - Case I



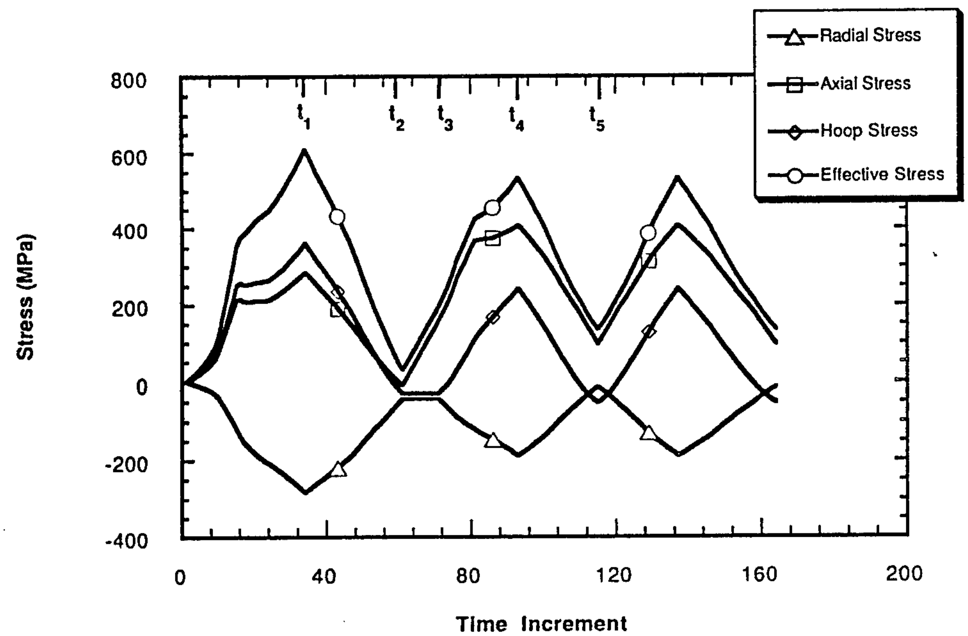
Matrix Stress History at Matrix/Fiber Interface - Case I



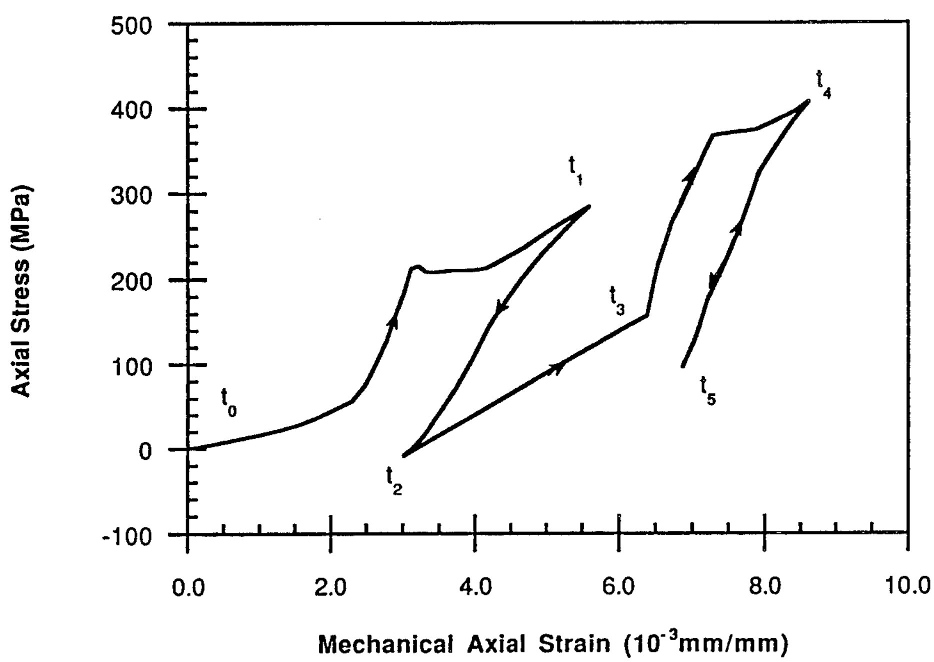
Thermal and Mechanical Loading - Case II



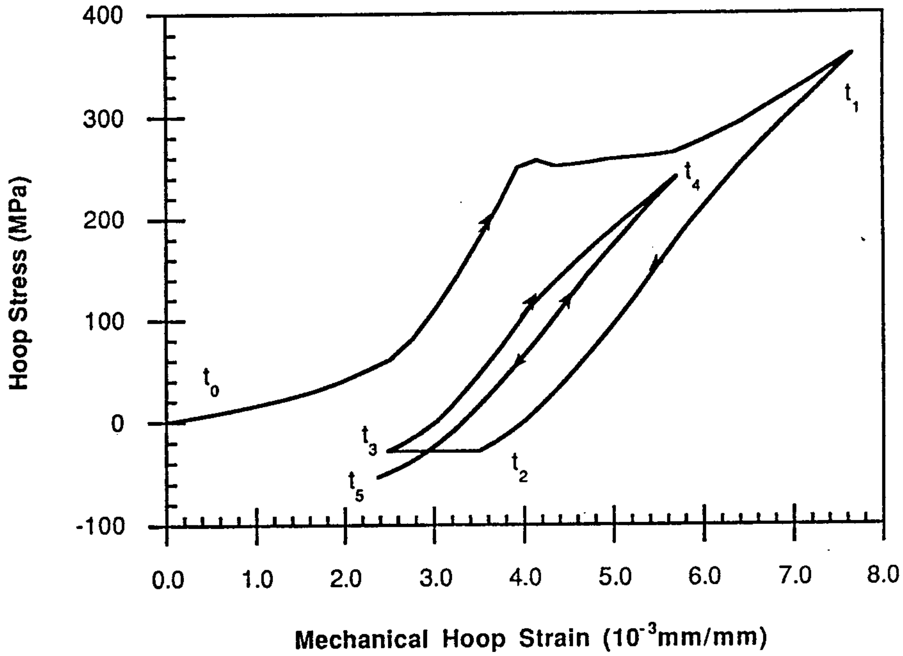
Matrix Stress History at Matrix/Fiber Interface - Case II



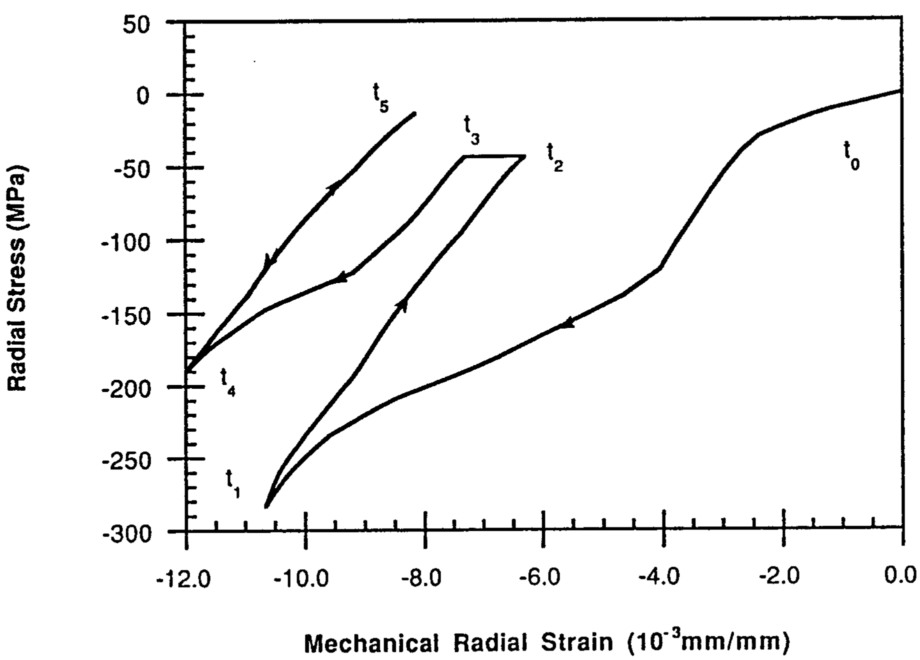
Matrix Axial Stress/Strain at Matrix/Fiber Interface
Case II



Matrix Hoop Stress/Strain at Matrix/Fiber Interface
Case II



Matrix Radial Stress/Strain at Matrix/Fiber Interface
Case II



Summary

- The use of a thermal elastic theory can predict the stresses above the material's yield stress during the MMC consolidation. A plasticity analyses is require to give a more realistic stress prediction.

- The material behavior of Titanium Aluminides is temperature dependent. To account for this behavior requires additional differential terms in the isothermal plasticity theory.

- This thermal elastic-plastic theory has been implemented into the finite element package MAGNA and is used to predict the behavior of MMCs subjected to thermal and mechanical loads.

- Simpler elastic analyses can been used to determine the stress ranges of MMC for these two cyclic load cases, but would not be able to predict the actual stress levels.

Future Activities

- Need to demonstrate simultaneous thermal and mechanical cyclic loading of MMC.

- Account for time dependent material behavior

  - creep model in the classical plasticity model

  (or)

  - unified inelastic strain model (Bodner/Partom etc.)

- Investigate monotonic and cyclic transverse loading of MMC

A VARIATIONAL APPROACH FOR DEVELOPMENT OF CONSISTENT CONSTITUTIVE RELATIONS FOR LAMINATED PLATES

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ABSTRACT

Constitutive relations for force resultant in plates are generally set up by direct integration of the stress-strain relations with appropriate assumptions regarding variation of displacement and generalized plane stress conditions. To properly allow for shear effects, shear correction factors have been used.

For application to laminates, shear correction factors appropriate to the specific applications, e.g., vibration, deformation, etc., have been proposed. However, no general procedure is available for selection of these factors for a given lay-up.

Hong [1,2] used an extension of Reissner's [3] method to set up consistent constitutive equations for shear resultants in laminated plates. However, his assumptions regarding displacement variation were rather simplistic. Also he did not address the need for constitutive relations for the other force resultants. He had assumed linear variation of the inplane displacement components over each layer and a transverse displacement independent of the transverse coordinate. Application of his theory to free-edge delamination specimens has shown that the accuracy of the solution increases with subdivision of each layer into sublayers. This has been the motivation for the development of a systematic approach to development of higher order theories based on higher order polynomial approximations for displacements.

In this paper, we start by stating a general variational equality based upon an extension of Reissner's work. Assuming a stress state satisfying equilibrium along with consistent assumptions regarding variation of displacement components over the thickness and the lateral coordinates of the laminated plate, constitutive equations for all the force resultants are derived. The number of these force resultants depends upon the order of the theory and includes interlayer tractions as well as the layerwise force resultants. In addition to obtaining expressions for layer deformations in terms of the force resultant, the procedure gives the relationships for interlayer slip and delamination opening in terms of the force resultants and the interface tractions. Thus, it is possible to incorporate the effects of damage directly into the field equations of the problem. The field equations including the equations of equilibrium, the constitutive equations and the continuity equations constitute a self-adjoint set and are amenable to Ritz type solution procedures including potential energy type finite element methods. Extension to complementary, mixed or hybrid procedures are direct.

ACKNOWLEDGEMENTS

The work reported is part of the work done at The Ohio State University under WRDC Grant F33625-85-C3213. Dr. George P. Sendeckyj is the Program Manager. Assistance provided by the Ohio State University Instruction and Research Computer Center and the Ohio Supercomputer Center is gratefully acknowledged.

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3. E. Reissner, "On a Certain Mixed Variational Theorem and a Proposed Application", I.J.N.M.E., Vol. 20, 1366-1368, 1984.

A VARIATIONAL APPROACH FOR DEVELOPMENT OF CONSISTENT CONSTITUTIVE RELATIONS FOR LAMINATED PLATES

by

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August 1990

APPROACH

- VARIATIONAL EQUALITY
- HIGHER ORDER DISCRETE LAMINATE THEORY
- CONSISTENT EQUILIBRIUM EQUATIONS
- EQUILIBRIUM STRESS STATE

OBJECTIVE

- CONSISTENT CONSTITUTIVE EQUATIONS
- INCLUSION OF INTERLAYER TRACTIONS
- CONSIDERATION OF DAMAGE

ASSUMED DISPLACEMENT THEORIES

$$u_\alpha = v_\alpha + x_3\phi_\alpha + \frac{x_3^2}{2}\psi_\alpha + \_\_ \text{ degree } n$$

$$u_3 = v_3 + x_3\phi_3 + \frac{x_3^2}{2}\psi_3 + \_\_ \text{ degree } m$$

	n=1, m=0	n=2, m=1
FIELD VARIABLES	5	8
EQUILIBRIUM EQNS.	5	8
FORCE RESULTANTS	8	14
CONSTITUTIVE EQNS.	8	14

FOR LAMINATES

MULTIPLY BY N

ADD INTERLAYER TRACTIONS 3(N-1)

ADD CONTINUITY CONDITIONS 3(N-1)

EQUILIBRIUM

$$\sigma_{ij,i} + f_j = 0$$

WEAK FORM

$$\int_{-\frac{t}{2}}^{\frac{t}{2}} (\sigma_{ij,i} + f_j) x_3^n dx_3 = 0 \,, \qquad n = 0, 1, \ldots$$

$n=0 \Rightarrow$

$$N_{\alpha\beta,\beta} + \sigma^+_{\alpha 3} - \sigma^-_{\alpha 3} = 0$$

$$Q_{\alpha,\alpha} + \sigma^+_{33} - \sigma^-_{33} = 0$$

$n=1 \Rightarrow$

$$M_{\alpha\beta,\beta} - Q_\alpha + \frac{t}{2}(\sigma^+_{\alpha 3} + \sigma^-_{\alpha 3}) = 0$$

$$M_{\alpha 3,\alpha} - N_{33} + \frac{t}{2}(\sigma^+_{33} + \sigma^-_{33}) = 0$$

$n=2 \Rightarrow$

$$P_{\alpha\beta,\beta} + \frac{t^2}{4}(\sigma^+_{\alpha 3} - \sigma^-_{\alpha 3}) - 2M_{\alpha 3} = 0$$

HONG'S FIRST ORDER THEORY

ASSUMPTIONS

DISPLACEMENTS

$$u^{(k)}_\alpha = v^{(k)}_\alpha(x_\beta) + x^{(k)}_3 \phi^{(k)}_\alpha(x_\beta)$$

$$u^{(k)}_3 = u_3(x_\beta)$$

SET OF STRESSES IN EQUILIBRIUM

$$\sigma^{(k)}_{\alpha\beta} = \frac{N^{(k)}_{\alpha\beta}}{t_k} + \frac{12x_3}{t_k^3} M^{(k)}_{\alpha\beta}$$

$$\sigma^{(k)}_{\alpha 3} = \zeta^{(k)}_1 Q^{(k)}_\alpha + \zeta^{(k)}_2 T^{(k-1)}_\alpha + \zeta^{(k)}_3 T^{(k)}_\alpha$$

A MIXED VARIATIONAL PRINCIPLE

FOR NO BODY FORCE

$$\Omega = \int_R \sigma_{ij}(u_{i,j} - \frac{1}{2}e_{ij})dR + \int_{S_2} u_i \hat{t}_i dS + \int_{S_1} T_i(u'_i - g_i)dS$$

GATEAUX DIFFERENTIAL = 0 =>

$$\int_A \{\sum_{k=1}^{N} \int_{-\frac{t_k}{2}}^{\frac{t_k}{2}} [\tau^{(k)}_{\alpha\beta}(u^{(k)}_{(\alpha,\beta)} - e^{(k)}_{\alpha\beta}) + \tau^{(k)}_{\alpha 3}(u^{(k)}_{\alpha,3} + u^{(k)}_{3,\alpha} - 2e^{(k)}_{\alpha 3})$$

$$+ \tau^{(k)}_{33}(u^{(k)}_{3,3} - e^{(k)}_{33})]dt + T^{(k)}_i(u'^{(k)}_i - g^{(k)}_i)\}dA = 0$$

HONG'S FIRST ORDER THEORY

2. CONSTITUTIVE RELATIONS

VARIATIONAL EQUALITY =>

$$u_{3,\alpha} + \phi^{(k)}_\alpha = 4C^{(k)}_{\alpha 3\beta 3}[\frac{6}{5t_k}Q^{(k)}_\beta - \frac{1}{10}(T^{(k-1)}_\beta + T^{(k)}_\beta)]$$

$$k = 1, 2, \ldots N$$

$$0 = C^{(k)}_{\alpha 3\beta 3}[3Q^{(k)}_\beta + t_k T^{(k-1)}_\beta - 4t_k T^{(k)}_\beta]$$

$$+ C^{(k)}_{\alpha 3\beta 3}[3Q^{(k+1)}_\beta + t_{k+1} T^{(k+1)}_\beta - 4t_{k+1} T^{(k)}_\beta]$$

CONDENSATION =>

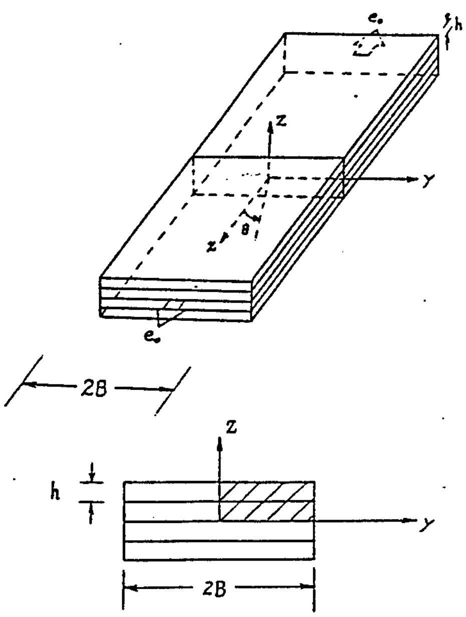
$$\phi^{(k)}_\alpha + u_{3,\alpha} = \sum_{j=1}^{N} \mu^{(k)j}_{\alpha\beta} Q^j_\beta$$

$$Q^{(k)}_\alpha = \sum_{j=1}^{N} \lambda^{(k)j}_{\alpha\beta}(\phi^j_\beta + u_{3,\beta})$$

HONG'S FIRST ORDER THEORY

3. APPLICATON TO FEDS

a. The Problem

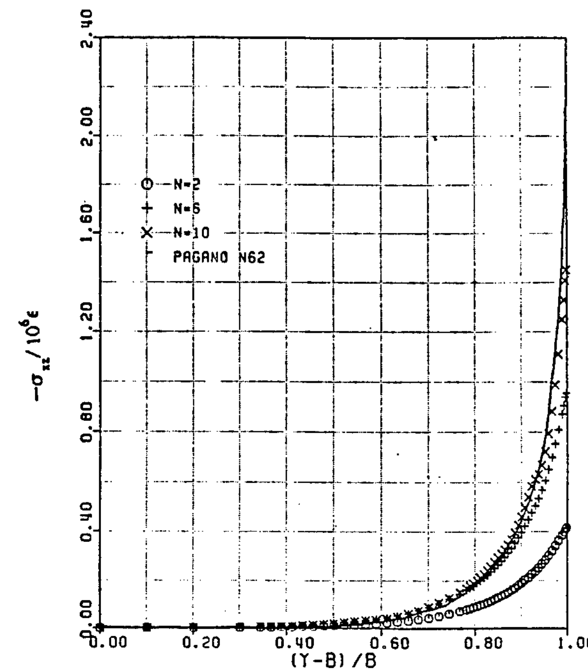


x = constant plane

HONG'S FIRST ORDER THEORY

3. APPLICATIONS

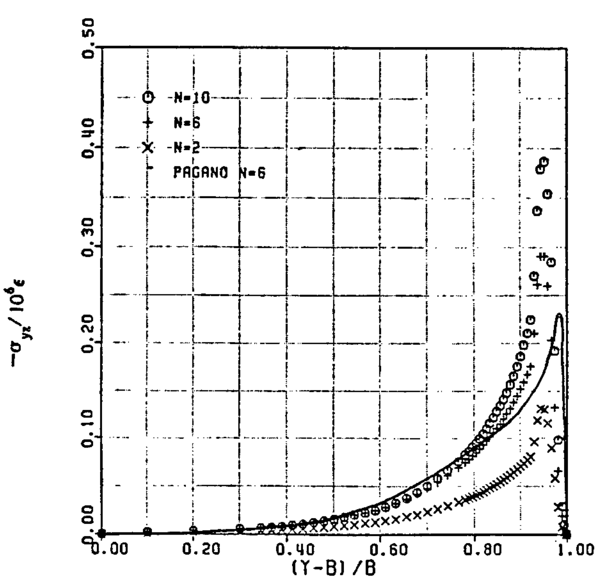
b. Distribution of σ_{xz} along ±45 interface.



HONG'S FIRST ORDER THEORY

3. APPLICATIONS

b. Distribution of σ_{yz} along 0/90 interface.



A HIGHER ORDER THEORY

1. KINEMATICS

DISPLACEMENT

$$u_\alpha = v_\alpha^{(k)}(x_\beta)+x_3^{(k)}\phi_\alpha^{(k)}(x_\beta)+\frac{1}{2}(x_3^{(k)})^2\psi_\alpha^{(k)}(x_\beta)$$

$$u_3^{(k)}=v_3^{(k)}(x_\beta)+x_3^{(k)}\phi_3^{(k)}(x_\beta)$$

STRAINS FROM DISPLACEMENTS

$$u_{(\alpha,\beta)}^{(k)}=v_{(\alpha,\beta)}^{(k)}+x_3^{(k)}\phi_{(\alpha,\beta)}^{(k)}+\frac{1}{2}x_3^{(k)2}\psi_{(\alpha,\beta)}^{(k)}$$

$$u_{\alpha,3}^{(k)}+u_{3,\alpha}^{(k)}=v_{3,\alpha}^{(k)}+\phi_\alpha^{(k)}+x_3^{(k)}(\phi_{3,\alpha}^{(k)}+\psi_\alpha^{(k)})$$

$$u_{3,3}^{(k)}=\phi_3^{(k)}$$

A HIGHER ORDER THEORY

2. STRESSES

$$\sigma_{\alpha\beta} = \eta_1^{(k)} N_{\alpha\beta}^{(k)} + \eta_2^{(k)} M_{\alpha\beta}^{(k)} + \eta_3^{(k)} P_{\alpha\beta}^{(k)}$$

$$\sigma_{\alpha 3}^{(k)} = \zeta_1^{(k)} Q_\alpha^{(k)} + \zeta_2^{(k)} M_{\alpha 3}^{(k)} + \zeta_3^{(k)} T_\alpha^{(k-1)} + \zeta_4^{(k)} T_\alpha^{(k)}$$

$$\sigma_{33}^{(k)} = \xi_1^{(k)} N_{33}^{(k)} + \xi_2^{(k)} T_3^{(k-1)} + \xi_3^{(k)} T_3^{(k)} + \xi_4^{(k)} T_{\alpha,\alpha}^{(k-1)} + \xi_5^{(k)} T_{\alpha,\alpha}^{(k)}$$

$$\psi_{(\alpha,\beta)}^{(k)} = \frac{15}{t_k^5} C_{\alpha\beta\gamma\delta}^{(k)} [-t_k^2 N_{\gamma\delta}^{(k)} + 12 P_{\gamma\delta}^{(k)}] + \frac{1}{14 t_k^3} [1120 N_{33}^{(k)}$$
$$+ 60 t_k (T_3^{(k-1)} + T_3^{(k)}) - 105 t_k^2 (T_{\rho,\rho}^{(k-1)} - T_{\rho,\rho}^{(k)})]$$

$$\phi_3^{(k)} = \frac{60}{7 t_k^3} C_{33\gamma\delta}^{(k)} [7 t_k^2 N_{\gamma\delta}^{(k)} - P_{\gamma\delta}^{(k)}]$$
$$+ \frac{1}{84 t_k} C_{3333}^{(k)} [120 N_{33}^{(k)} - 18 t_k (T_3^{(k-1)} + T_3^{(k)})$$
$$+ t_k^{(2)} (T_{\rho,\rho}^{(k-1)} - T_{\rho,\rho}^{(k)})]$$

VARIATONAL EQUALITY =>

$$\phi_\alpha^{(k)} + u_{3,\alpha}^{(k)} = \frac{2}{5} C_{\alpha 3\gamma 3}^{(k)} [\frac{12}{t_k} Q_\gamma^{(k)} - (T_\gamma^{(k-1)} + T_\gamma^{(k)})]$$

$$2\psi_\alpha^{(k)} + \phi_{3,\alpha}^{(k)} = \frac{6}{7 t_k} C_{\alpha 3\gamma 3}^{(k)} [\frac{80}{t_k} M_{\gamma 3}^{(k)} + T_\gamma^{(k-1)} - T_\gamma^{(k)}]$$

$$v_{(\alpha,\beta)}^{(k)} = \frac{1}{4 t_k^3} C_{\alpha\beta\gamma\delta}^{(k)} [t_k^2 N_{\gamma\delta}^{(k)} - 60 P_{\alpha\beta}^{(k)}] + \frac{1}{56 t_k} C_{\alpha\beta 33}^{(k)}$$
$$[3360 N_{33}^{(k)} - 140 t_k (T_3^{(k-1)} + t_3^{(k)} + t_k^2 (T_{\rho,\rho}^{(k-1)} - T_{\rho,\rho}^{(k)})]$$

$$\phi_{(\alpha,\beta)}^{(k)} = \frac{12}{t_k^3} C_{\alpha\beta\gamma\delta}^{(k)} M_{\gamma\delta}^{(k)} + \frac{1}{10 t_k} C_{\alpha\beta 33}^{(k)} [-12 (T_3^{(k-1)} - T_3^{(k)})$$
$$+ t_k (T_{\rho,\rho}^{(k-1)} - T_{\rho,\rho}^{(k)})]$$

$$0 = C_{33\gamma\delta}^{(k)} [-\frac{5}{2} N_{\gamma\delta}^{(k)} + \frac{6}{5 t_k} M_{\gamma\delta}^{(k)} + \frac{30}{7 t_k^2} P_{\gamma\delta}^{(k)}]$$
$$+ C_{33\gamma\delta}^{(k+1)} [-\frac{5}{2} N_{\gamma\delta}^{(k+1)} - \frac{6}{5 t_{k+1}} M_{\gamma\delta}^{(k+1)} + \frac{30}{7 t_{k+1}^2} P_{\gamma\delta}^{(k+1)}]$$
$$+ C_{3333}^{(k)} [-\frac{3}{14} N_{33}^{(k)} - \frac{t_k}{70} T_3^{(k-1)} + \frac{8 t_k}{35} T_3^{(k)} + \frac{t_k^2}{210} T_{\rho,\rho}^{(k-1)} + \frac{t_k^2}{60} T_{\rho,\rho}^{(k)}]$$
$$+ C_{3333}^{(k)} [-\frac{3}{14} N_{33}^{(k+1)} + \frac{8 t_{k+1}}{35} T_3^{(k)} - \frac{t_{k+1}}{70} T_3^{(k+1)} - \frac{t_k^2}{60} T_{\rho,\rho}^{(k)} - \frac{t_k^2}{210} T_{\rho,\rho}^{(k+1)}]$$

$$0 = C_{\alpha 3\gamma 3}^{(k)} [-1008 Q_\alpha^{(k)} - \frac{2160}{t_k} M_{\gamma 3}^{(k)} + 144 t_k T_\gamma^{(k-1)} + 864 t_k T_\gamma^{(k)}]$$
$$+ C_{\alpha 3\gamma 3}^{(k+1)} [-1008 Q_\gamma^{(k+1)} + \frac{2160}{t_{k+1}} M_{\gamma 3}^{(k+1)} + 864 t_{k+1} T_\gamma^{(k)} + 144 t_{k+1} T_\gamma^{(k+1)}]$$
$$+ C_{33\gamma\delta}^{(k)} [45 N_{\gamma\delta,\alpha}^{(k)} - \frac{252}{t_k} M_{\gamma\delta,\alpha}^{(k)} - 18900 P_{\gamma\delta,\alpha}^{(k)}]$$
$$+ C_{33\gamma\delta}^{(k+1)} [-45 N_{\gamma\delta,\alpha}^{(k+1)} - \frac{252}{t_k} M_{\gamma\delta,\alpha}^{(k+1)} + 18900 P_{\gamma\delta,\alpha}^{(k+1)}]$$
$$+ C_{3333}^{(k)} [30 t_k N_{33,\alpha}^{(k)} + 12 t_k^2 T_{3,\alpha}^{(k-1)} - 42 t_k^2 T_{3,\alpha}^{(k)} - 2 t_k^3 T_{\rho,\rho\alpha}^{(k-1)} - 4 t_k^3 T_{\rho,\rho\alpha}^{(k)}]$$
$$+ C_{3333}^{(k+1)} [-30 t_{k+1} N_{33,\alpha}^{(k+1)} + 42 t_{k+1}^2 T_{3,\alpha}^{(k)} - 12 t_{k+1}^2 T_{3,\alpha}^{(k+1)}$$
$$-4 t_{k+1}^3 T_{\rho,\rho\alpha}^{(k)} - 2 t_{k+1}^3 T_{\rho,\rho\alpha}^{(k+1)}]$$

APPENDIX A: PROGRAM LISTINGS

NASA LANGLEY RESEARCH CENTER

IN-HOUSE

ADVANCED CONCEPTS FOR COMPOSITE HELICOPTER FUSELAGE STRUCTURES
83 April 1 - 92 January 1

Project Engineer: Mr. Donald J. Baker
Mail Stop 190
Aerostructures Directorate, USAARTA (AVSCOM)
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3171 FTS 928-3171

Objective: To investigate new design concepts for composite materials on lightly loaded helicopter fuselage structures. Trade studies will be performed using the various computer codes. A 4-year task assignment contract was awarded in Fiscal Year 1989 to fabricate selected designs that will be tested at NASA Langley.

POSTBUCKLING AND CRIPPLING OF COMPRESSION-LOADED COMPOSITE STRUCTURAL COMPONENTS
79 March 1 - 91 September 30

Project Engineer: Dr. James H. Starnes, Jr.
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3168 FTS 928-3168

Objective: To study the postbuckling and crippling of compression-loaded composite components and to determine the limitations of postbuckling design concepts in structural applications.

DESIGN TECHNOLOGY FOR STIFFENED CURVED COMPOSITE PANELS AND SHELLS
79 October 1 - 91 September 30

Project Engineer: Dr. James H. Starnes, Jr.
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3168 FTS 928-3168

Objective: To develop verified design technology for generic advanced-composite stiffened curved panels.

POSTBUCKLING OF FLAT STIFFENED GRAPHITE/EPOXY SHEAR WEBS
81 July 1 - 91 September 30

Project Engineer: Mr. Marshall Rouse
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3182 FTS 928-3182

Objective: To study the postbuckling response and failure characteristics of flat stiffened graphite/epoxy shear webs.

ASH CHARACTERISTICS OF COMPOSITE FUSELAGE STRUCTURES
July 1 - 91 September 30

oject Engineer: Mr. Huey D. Carden
Mail Stop 495
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-4151 FTS 928-4151

ojective: To study the crash characteristics of composite transport fuselage structural components.

JCKLING AND STRENGTH OF THICK-WALLED COMPOSITE CYLINDERS
October 1 - 91 September 30

oject Engineer: Ms. Dawn C. Jegley
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3185 FTS 928-3185

bjective: To develop accurate analyses for the buckling and strength predictions of thick-walled composite cylinders.

DVANCED COMPOSITE STRUCTURAL CONCEPTS
4 October 1 - 92 September 30

roject Engineer: Dr. James H. Starnes, Jr.
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3168 FTS 928-3168

bjective: To develop composite structural concepts and design technology needed to realized the improved performance, structural efficiency, and lower-cost advantage offered by new material systems and manufacturing methods for advanced aircraft structures.

AILURE MECHANISMS FOR COMPOSITE LAMINATES WITH DAMAGE AND LOCAL ISCONTINUITIES
6 October 1 - 91 September 30

roject Engineer: Dr. Mark J. Shuart
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3170 FTS 928-3170

Objective: To study the effects of impact damage and local discontinuities on the strength of composite structural components, to identify the failure modes that govern the behavior of components subjected to low-velocity impact damage, and to analytically predict failure and structural response.

MECHANICS OF ANISOTROPIC COMPOSITE STRUCTURES
86 October 1 - 91 September 30

Project Engineer: Dr. Michael P. Nemeth
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3184 FTS 928-3184

Objective: To develop analytical procedures for anisotropic structural components that accurately predict the response for tailored structures.

QUANTITATIVE ACOUSTIC EMISSION ANALYSIS OF ADHESIVE BOND FAILURE

Project Engineer: Mr. William H. Prosser
IRD, Nondestructive Measurement Science Branch
Mail Stop 231
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-4960 FTS 928-4960

Objective: The objective is to study the influence of fracture toughness of the adhesive, mode of fracture, and crack velocity on the acoustic emission released during adhesive bond failure.

EXPERIMENTAL AND ANALYTICAL CHARACTERIZATION OF THE MECHANICAL BEHAVIOR OF METAL MATRIX COMPOSITES
80 June - 91 September 30

Project Engineer: Dr. W. Steven Johnson
Mail Stop 188E
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3463 FTS 928-3463

Objective: To experimentally investigate the fatigue, fracture, and thermomechanical behavior of MMC's to insure airframe structural integrity at elevated temperatures. Both continuously reinforced laminates and discontinuous particulate and whisker reinforced MMC's will be included in the study.

DEVELOPMENT OF ANALYTICAL MODELS OF THE THERMOMECHANICAL BEHAVIOR OF METAL MATRIX COMPOSITES
87 June - 91 September 30

Project Engineer: Dr. C. A. Bigelow
Mail Stop 188E
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3462 FTS 928-3462

Objective: To develop finite-element codes, laminate-analysis codes, and micromechanics models necessary to analytically investigate mechanics issues related to the fatigue, fracture, and thermomechanical behavior of MMC's.

DELAMINATION MICROMECHANICS ANALYSIS
85 October 1 - 91 September 30

Project Engineer: Dr. John H. Crews, Jr.
Mail Stop 188E
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3457 FTS 928-3457

Objective: To develop stress analysis for region near a delamination front, including microcracks and fiber bridging.

MECHANICS MODELS OF ADVANCED TEXTILE COMPOSITES
88 June 1 - 91 September 30

Project Engineer: Mr. Clarence C. Poe, Jr.
Mail Stop 188E
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3449 FTS 928-3449

Objective: To develop finite-element models of the deformation and local stress states that reflect the local fiber curvature of advanced textile composites. Mathematical descriptions of the unit cell architecture will be the basis for the models. Failure criteria will be developed to optimize these materials with regard to in-plane and out-of-plane strength. Experiments will be conducted to support model development and verify predictions.

INTERLAMINAR SHEAR FRACTURE TOUGHNESS
89 May - 91 September 30

Project Engineer: Ms. Gretchen B. Murri
Mail Stop 188E
Aerostructures Directorate, USAARTA (AVSCOM)
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3466 FTS 928-3466

Objective: Cyclic end-notched flexure tests will be used to measure the mode II strain energy release rate of two materials in fatigue. Results will be used to develop ASTM test standards for strain energy release rate under fatigue loading.

DELAMINATIONS IN TAPERED COMPOSITE LAMINATES WITH INTERNAL PLY DROPS
88 October - 90 September 30

Project Engineer: Ms. Gretchen B. Murri
Mail Stop 188E
Aerostructures Directorate, USAARTA (AVSCOM)
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3466 FTS 928-3466

Objective: To characterize delamination failures in tapered composites containing internal ply-drops. Experimental results from a variety of materials and lay-ups will be compared with finite element and closed-form solutions.

STUDY OF DAMAGE TOLERANCE OF THERMOPLASTIC COMPOSITES
88 December - 92 December

Project Engineer: Mr. C. C. Poe, Jr.
Mail Stop 188E
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3467 FTS 928-3467

Objective: To develop an analysis that can also be used as a design tool for predicting the complete damage state during and after impact and the residual properties in thermoplastic and thermoset matrices.

IMPACT RESPONSE AND DAMAGE IN THREE-DIMENSIONAL BRAIDED GRAPHITE FIBER COMPOSITES
87 October - 90 October

Project Engineer: Mr. C. C. Poe, Jr.
Mail Stop 188E
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3467 FTS 928-3467

Objective: To characterize damage in three-dimensional braided composites subjected to hard object impact at low energy levels.

MIXED-MODE DELAMINATION TESTING
87 September 1 - 90 September 30

Project Engineer: Mr. James R. Reeder
Mail Stop 188E
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3456 FTS 928-3456

Objective: To measure the delamination fracture toughness of laminated composites material subjected to combined mode I and mode II loadings and thereby develop a mixed mode delamination failure criterion. The new mixed-mode-bending specimen will be used for testing.

HIGH TEMPERATURE LONG-TERM APPLICATIONS OF POLYMERIC COMPOSITES
90 January 1 - 98 September 30

Project Engineer: Dr. W. S. Johnson
Mail Stop 188E
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3463

Objective: To assess the influence of thermal-mechanical fatigue and long term durability on the mechanical properties of polymeric matrix composites for use on advanced supersonic commercial transports. Temperatures will approach 450°F for 60,000 flight hours.

TIME DEPENDENT COMPOSITE CHARACTERIZATION FOR POLYMER COMPOSITES
90 January 1 - 95 September 30

Project Engineer: Dr. Thomas S. Gates
Mail Stop 188E
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3400 FTS 928-3400

Objective: To develop constitutive models for nonlinear, rate dependent behavior. The analysis will be supported by experimental data to account for creep, relaxation, and physical aging.

EXPERIMENTAL EVALUATION OF ADVANCED COMPOSITE MATERIAL FORMS
84 June 1 - 91 June 1

Project Engineer: Mr. H. Benson Dexter
Mail Stop 188B
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3094 FTS 928-3094

Objective: To determine mechanical properties and establish damage tolerance of 2-D and 3-D woven, stitched, and braided composite materials.

FLIGHT SERVICE EVALUATION OF COMPOSITE COMPONENTS ON COMMERCIAL AND MILITARY AIRCRAFT
72 March 1 - 90 December 31

Project Engineer: Mr. H. Benson Dexter
Mail Stop 188B
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3094 FTS 928-3094

Objective: To evaluate the long-term durability of composite components installed on commercial and military transport and helicopter aircraft. Over 300 components constructed of boron, graphite, and Kevlar composites will be evaluated after extended service. Components include graphite/epoxy rudders, spoilers, tail rotors, vertical stabilizers, Kevlar/epoxy fairings, doors and ramp skins, and boron/aluminum aft pylon skins. Note: Over 5.0 million total component flight hours have been accumulated since initiation of flight service in 1972. Composite components on L-1011, B-737, and DC-10 aircraft have accumulated over 40,000 flight hours each. Excellent in-service performance and maintenance experience has been achieved with the composite components.

MICROMECHANICS MODELING OF COMPOSITE THERMOELASTIC BEHAVIOR
86 October - 91 June 30

Project Engineer: Dr. David E. Bowles
Mail Stop 188B
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3095 FTS 928-3095

Objective: Develop analytical methods to investigate thermally induced deformations and stresses in continuous fiber-reinforced composites at the micromechanics level, and predict how these deformations and stresses affect the dimensional stability of the composite.

THERMAL DEFORMATIONS AND STRESSES IN COMPOSITE/HONEYCOMB PANELS FOR PRECISION REFLECTORS
89 June 1 - 91 May 31

Project Engineer: Dr. David E. Bowles
Mail Stop 188B
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3095 FTS 928-3095

Objective: Analytically and experimentally investigate the effects of constituent properties (fiber, matrix, honeycomb, adhesive) on thermally induced deformations and stresses in composite honeycomb panels for precision reflector applications.

ADVANCED COMPOSITE MATERIALS FOR ULTRA-HIGH PRECISION REFLECTOR HONEYCOMB PANELS
88 October 1 - 91 September 30

Project Engineer: Dr. Stephen S. Tompkins
Mail Stop 188B
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3096 FTS 928-3096

Objective: Develop advanced structural composite materials that are dimensionally stable and durable in the LEO and GEO space environments. These materials will form facesheets of honeycomb panels used to construct precision reflector panels for space applications. A critical requirement for the facesheet materials is to replicate a highly polished, very accurate mold surface (surface accuracy about 3 microns RMS).

THERMAL AND MECHANICAL STABILITY OF COMPOSITE MATERIALS
83 October 1 - 93 September 30

Project Engineer: Dr. Stephen S. Tompkins
Mail Stop 188B
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3096 FTS 928-3096

Objective: Develop and evaluate structural composite materials (resin-, metal-, and glass-matrix) that are dimensionally stable and/or have stable thermal and mechanical properties when subjected to simulated long-term LEO and GEO space service environments.

CONTRACTS

COLLAPSE AND FAILURE MODES IN ADVANCED COMPOSITE STRUCTURES
NSG-1483
78 January 15 - 90 January 14

Project Engineer: Dr. James H. Starnes, Jr.
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3168 FTS 928-3168

Principal Investigator: Dr. Wolfgang G. Knauss
California Institute of Technology
Pasadena, CA 91125
(213) 356-4524/4528

Objective: To experimentally and analytically study time-dependent effects on buckling and failure of composite structures with discontinuities.

ADVANCED COMPOSITE STRUCTURAL DESIGN TECHNOLOGY FOR COMMERCIAL TRANSPORT AIRCRAFT
Pending
90 December 1 - 96 December 1

Project Engineer: Dr. James H.Starnes, Jr.
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3168 FTS 928-3168

Principal Investigator: TBD

Objective: To design, analyze, fabricate, and test generic advanced-composite structural components for subsonic and supersonic transport aircraft applications in order to develop verified design technology.

STRUCTURAL OPTIMIZATION FOR IMPROVED DAMAGE TOLERANCE
NAG-1-168
81 September 1 - 91 October 15

Project Engineer: Dr. James H. Starnes, Jr.
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3168 FTS 928-3168

Principal Investigator: Dr. Raphael T. Haftka
Virginia Polytechnic Institute and State University
Blacksburg, VA 24061
(703) 231-4860

Objective: To develop a structural optimization procedure for composite wing boxes that includes the influence of damage-tolerance considerations in the design process.

FAILURE ANALYSIS AND DAMAGE TOLERANCE OF COMPOSITE AIRCRAFT STRUCTURES
NAS1-17925
85 February 23 - 90 December 30

Project Engineer: Dr. Damodar R. Ambur
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3174 FTS 928-3174

Principal Investigator: Dr. R. K. Kunz
Lockheed Aeronautical Systems Co.
Burbank, CA 91520
(818) 847-7995

Objective: To develop advanced structural concepts and to advance the analytical capability to predict composite structural failure.

ANISOTROPIC SHELL ANALYSIS
NAG-1-901
88 October 1 - 91 September 30

Project Engineer: Dr. Michael P. Nemeth
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3184 FTS 928-3184

Principal Investigator: Dr. Michael W. Hyer
Virginia Polytechnic Institute and State University
Blacksburg, VA 24061
(703) 231-5372

Objective: To develop accurate analyses for the response of anisotropic composite shell structures.

THICKNESS DISCONTINUITY EFFECTS
NAG-1-537
85 October 1 - 91 September 30

Project Engineer: Dr. James H. Starnes, Jr.
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3168 FTS 928-3168

Principal Investigator: Dr. Eric R. Johnson
Virginia Polytechnic Institute and State University
Blacksburg, VA 24061
(703) 231-6126

Objective: To develop verified analytical models of compression loaded laminates with thickness discontinuities and dropped plies.

MECHANICS OF ANISOTROPIC STRUCTURES WITH CUTOUTS
NAG-1-917
88 December 1 - 91 December 1

Project Engineer: Dr. Michael P. Nemeth
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3184 FTS 928-3184

Principal Investigator: Dr. E. C. Klang
North Carolina State University
Raleigh, NC 27695
(919) 737-2365

Objective: To develop efficient analytical procedures that accurately predict the response of anisotropic structural components with cutouts.

STRUCTURAL DESIGN CRITERIA FOR FILAMENT-WOUND COMPOSITE SHELLS
NAG-1-982
89 May 15 - 91 May 15

Project Engineer: Dr. James H. Starnes, Jr.
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3168 FTS 928-3168

Principal Investigator: Dr. H. T. Hahn
Pennsylvania State University
University Park, PA 16802
(814) 865-4523

Objective: To develop structural design criteria that can be used to scale-up filament wound composite shells.

COMPOSITE FUSELAGE TECHNOLOGY
NAG-1-982
89 April 7 - 91 April 7

Project Engineer: Dr. James H. Starnes, Jr.
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3168 FTS 928-3168

Principal Investigator: Dr. P. A. Lagace
Massachusetts Institute of Technology
Cambridge, MA 02139
(617) 253-3628

Objective: To conduct experimental and analytical studies of pressurized composite fuselage shells subjected to damage.

FIBER BUCKLING IN LAMINATED PLATES
NAG-1-1040
89 October 1 - 91 September 30

Project Engineer: Dr. Mark J. Shuart
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3170 FTS 928-3170

Principal Investigator: Dr. A. Waas
University of Michigan
Ann Arbor, MI 48109-1248
(313) 764-8227

Objective: Conduct experimental and analytical studies to isolate and observe in-situ failure mechanisms for composite structures.

STIFFNESS TAILORING OF COMPOSITE PLATES FOR IMPROVED STABILITY AND STRENGTH UNDER COMBINED LOADING
NAG-1-1141
90 June 1 - 91 May 30

Project Engineer: Dr. Mark J. Shuart
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3170 FTS 928-3170

Principal Investigator: Dr. S. B. Biggers
Clemson University
Clemson, SC 29634
(803) 656-0139

Objective: Conduct experimental and analytical studies to tailor membrane and bending stiffnesses for a composite plate that will result in improved buckling resistance and/or postbuckling strength.

ADVANCED COMPOSITE STRUCTURAL CONCEPTS AND MATERIAL TECHNOLOGIES FOR PRIMARY AIRCRAFT STRUCTURES
NAS1-18888
1989 April - 1995 May

Project Engineer: Dr. Mark J. Shuart
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3170 FTS 928-3170

Principal Investigator: Mr. A. Jackson
Lockheed Aeronautical Systems Company
Department 7007
Building 369, Plant B6
Burbank, CA 91520
(818) 847-5450

Objective: To develop and verify innovative structural concepts such as geodesic stiffening, sandwich construction, and pultruded stiffeners that exploit the full potential of integrated design/manufacturing procedures to achieve light-weight and cost-effective primary structures; and to develop a strong structural mechanics technology base to predict the performance of advanced concepts.

FIBER WAVEGUIDE SENSORS FOR INTELLIGENT MATERIALS
NAG-1-895
1988 September - 1990 October

Project Engineer: Dr. Robert S. Rogowski
IRD, Nondestructive Measurement Science Branch
Mail Stop 231
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-4990 FTS 928-4990

Principal Investigator: Richard O. Claus
Department of Electrical Engineering
Virginia Polytechnic Institute and State University
Blacksburg, VA 24061

Objective: Development of fiber-optic based opto-electronic sensing instrumentation for the characterization of materials and structures.

QUANTITATIVE NONDESTRUCTIVE EVALUATION OF COMPOSITE MATERIALS BASED ON ULTRASONIC WAVE PROPAGATION
NSG-1-601
1981 September - 1990 September

Project Engineer: Dr. Eric Madaras
IRD, Nondestructive Measurement Science Branch
Mail Stop 231
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-4993 FTS 928-4993

Principal Investigator: Dr. James G. Miller
Department of Physics
Washington University
St. Louis, MO 33130

Objective: The overall goal of our research program is the development and application of quantitative ultrasonic techniques to problems of nondestructive evaluation of composites materials. One goal of this work is to demonstrate the potential application of approaches base on the relationship between frequency dependent attenuation and dispersion to nondestructive evaluation of porosity. A second goal is the use of quantitative polar backscatter and attenuation measurements to characterize material properties.

INVESTIGATION OF ACOUSTIC PROPERTIES OF COMPOSITE MATERIALS
NAG-1-431
1983 September - 1990 September

Project Engineer: Dr. Eric Madaras
IRD, Nondestructive Measurement Science Branch
Mail Stop 231
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-4993 FTS 928-4993

Principal Investigator: Dr. Barry T. Smith
Department of Physics
Christopher Newport College
Newport News, VA 23606

Objective: The research involves an investigation of the ultrasonic acoustic properties of composite materials. The objective is to characterize the material as well as develop means of assessing any damage. Research to date has included quantitative measurement of impact damage in thin graphite/epoxy composites, evaluation of porosity and determination of fundamental ultrasonic properties to elucidate the propagation of ultrasonic waves in these materials.

NONDESTRUCTION EVALUATION OF CARBON-CARBON COMPOSITES
1989 September - 1995 September

Project Engineer: Dr. Eric Madaras
IRD, Nondestructive Measurement Science Branch
Mail Stop 231
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-4993 FTS 928-4993

Principal Investigator: Dr. Ron Kline
Aerospace and Mechanical Engineering
University of Oklahoma
Norman, OK 73019

Objective: The research involves methods of measuring the elastic moduli of carbon-carbon material and integrating the results with FEM codes to predict the behavior of components. Also, research related to nondestructive evaluation of carbon-carbon coatings will be investigated.

PROGRESSIVE FAILURE MODEL FOR LAMINATED COMPOSITES
NAG-1-979
89 March1 - 92 February 28

Project Engineer: Dr. Charles E. Harris
Mail Stop 188E
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3449 FTS 928-3449

Principal Investigator: Dr. David H. Allen
Aerospace Engineering Department
Texas A&M University
College Station, TX 77843

Objective: To develop a damage-dependent constitutive model as the mechanics foundation for a progressive failure methodology to predict the residual strength and life of laminates.

THERMAL VISCOPLASTICITY IN METAL MATRIX COMPOSITES
L-24457C
87 July - 91 January

Project Engineer: Dr. W. S. Johnson
Mail Stop 188E
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3463 FTS 928-3463

Principal Investigator: Dr. Yehia A. Bahei-El-Din
Department of Civil Engineering
Rensselaer Polytechnic Institute
Troy, NY 12180-3590
(518) 276-8043

Objective: This contract is to develop an analytical method for estimating thermal viscoplasticity stresses and strains in continuous fiber-reinforced metal matrix composites due to fabrication and/or subsequent thermal cycling and mechanical loadings.

ANALYSIS OF INTERLAMINAR FRACTURE IN COMPOSITES UNDER COMBINED LOADING
NAG-1-637
89 October 1 - 90 September 30

Project Engineer: Ms. Gretchen B. Murri
Aerostructures Directorate, USAARTA (AVSCOM)
NASA Langley Research Center
Mail Stop 188E
Hampton, VA 23665-5225
(804) 864-3466 FTS 928-3466

Principal Investigator: Dr. E. A. Armanios
School of Aerospace Engineering
Georgia Institute of Technology
Atlanta, GA 30332

Objective: The objective of this program is to extend an existing sublaminate analysis method to model tapered ply-drop configurations under bending and combined tension-bending loads. The analyses are intended for use on personal-class computers.

DEVELOPMENT OF ADVANCED WOVEN COMPOSITE MATERIALS AND STRUCTURAL FORMS
NAS1-18358
86 August 29 - 91 August 29

Project Engineer: Mr. H. Benson Dexter
Mail Stop 188B
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3094 FTS 928-3094

Principal Investigator: Ms. Janice Maiden
Textile Technologies, Inc.
2800 Turnpike Drive
Hatboro, PA 19040
(215) 443-5325

Objective: To develop textile technology to produce 2-D and 3-D woven preforms and structural elements with integral stiffening, multilayers, and multidirectional reinforcement.

ANALYSIS OF 2-D AND 3-D REINFORCED COMPOSITES
ENVIRONMENTAL EXPOSURE EFFECT ON COMPOSITE MATERIALS FOR COMMERCIAL AIRCRAFT
NAS1-15148
77 November 1 - 90 December 31

Project Engineer: Mr. H. Benson Dexter
Mail Stop 188B
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3094 FTS 928-3094

Principal Investigator: Mr. Randy Coggeshall
Boeing Commercial Airplane Company
P.O. Box 3707
Seattle, WA 98124
(206) 234-6695

Objective: To provide technology in the area of environmental effects on graphite/epoxy composite materials, including long-term performance of advanced resin-matrix composite materials in ground and flight environments.

MECHANICAL PROPERTIES OF 3-D WOVEN FABRIC
NCC-1-130
88 August 1 - 91 May 1

Project Engineer: Dr. Gary L. Farley
Mail Stop 188B
Aerostructures Directorate, USAARTA (AVSCOM)
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3091 FTS 928-3091

Principal Investigator: Dr. John M. Kennedy
Department of Mechanical Engineering
Clemson University
Clemson, SC
(803) 656-5632

Objective: Establish the mechanical response and damage tolerance characteristics of 3-D woven fabrics.

VISCOELASTIC RESPONSE OF COMPOSITE/HONEYCOMB PANELS FOR PRECISION REFLECTORS
NAG-1-343
88 August 16 - 91 December 31

Project Engineer: Dr. D. E. Bowles
Mail Stop 188B
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3095 FTS 928-3095

Principal Investigator: Dr. M. W. Hyer
Virginia Polytechnic Institute and State University
Blacksburg, VA 24061
(703) 961-5372

Objective: Analytically and experimentally investigate the viscoelastic response of sandwich panels fabricated from composite facesheets and honeycomb cores.

UNIT CELL GEOMETRY OF COMPLEX PREFORMS FOR STRUCTURAL COMPOSITES
NCC-1-138
89 May 1 - 92 May 1

Project Engineer: Mr. H. Benson Dexter
Mail Stop 188B
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3094 FTS 928-3094

Principal Investigator: Drs. Christopher M. Pastore and Frank K. Ko
Department of Materials Engineering
Drexel University
Philadelphia, PA 19104
(215) 895-1844

Objective: Develop computerized graphics models for a variety of 2-D and 3-D textile fiber architectures for use in micromechanics analysis. Unit cells will be defined and a library of these cells can be used in fabric analysis or finite element models for stress analysis.

ADVANCED COMPOSITE FABRICATION AND TESTING
NAS1-18954
89 August - 94 August

Project Engineer: Mr. Marvin B. Dow
Mail Stop 188B
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3090 FTS 928-3090

Principal Investigator: Mr. Anthony Falcone
Boeing Aerospace
Seattle, WA 98124
(206) 234-2678

Objective: To process and test experimental composite materials and state-of-the-art systems including woven, braided, knitted, and stitched fiber forms. Processing shall include resin transfer molding, pultrusion, and thermoforming.

DEVELOPMENT OF FILAMENT WINDING PROCESS FOR GR/TP COMPOSITE LAMINATES
NAS1-18624
89 April 27 - 90 July 27

Project Engineer: Mr. Jerry W. Deaton
Mail Stop 188B
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3087 FTS 928-3087

Principal Investigator: Mr. G. E. Walker, Jr.
Hercules Aerospace Company
Composite Products Group
Bacchus Works
Magna, UT 84044-0098
(801) 251-4194

Objective: Development of consolidation processes for Gr/TP filament-wound/fiber-placement laminates and demonstration of laminate quality through nondestructive evaluation/inspection and mechanical testing. Machining of specimens from Gr/TP laminates and all mechanical testing will be accomplished at NASA Langley.

ADVANCED COMPOSITE STRUCTURAL CONCEPTS AND MATERIAL TECHNOLOGIES FOR PRIMARY AIRCRAFT STRUCTURES
NAS1-18888
1989 April - 1995 May

Project Engineer: Dr. Mark J. Shuart
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3170 FTS 928-3170

Principal Investigator: Mr. A. Jackson
Lockheed Aeronautical Systems Company
Department 7007
Building 369, Plant B6
Burbank, CA 91520
(818) 847-5450

Objective: To develop and verify innovative structural concepts such as geodesic stiffening, sandwich construction, and pultruded stiffeners that exploit the full potential of integrated design/manufacturing procedures to achieve light-weight and cost-effective primary structures; and to develop a strong structural mechanics technology base to predict the performance of advanced concepts.

NOVEL MATRIX RESINS WITH IMPROVED PROCESSABILITY AND PROPERTIES FOR PRIMARY AIRCRAFT STRUCTURES
NAS1-18841
1989 April - 1992 April

Project Engineer: Dr. P. Hergenrother
Mail Stop 226
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-4270 FTS 928-4270

Principal Investigator: Dr. E. P. Woo
Dow Chemical Company
1712 Building
Midland, MI 48674
(517)-636-1072

Objective: Develop new high performance thermoset resins with improved durability, toughness and processability. The resins will be targeted for aircraft structural applications with maximum use temperatures ranging from 180°F to 450°F. New resins including vinyl esters, cyanates, modified epoxies, acetylene-terminated polymers and bisbenzocyclobutenes will be synthesized. The suitability of the new resin technology for use in resin transfer molding as well as for conventional prepreg processing will be evaluated.

ADVANCED MATERIALS AND PRODUCT FORMS
NAS1-18834
1989 April - 1995 May

Project Engineer: Dr. N. Johnston
Mail Stop 188M
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3493 FTS 928-3493

Principal Investigator: Mr. J. T. Hartness
BASF
13504-A South Point Blvd.
Charlotte, NC 28217
(704)-588-7976

Objective: Develop improved matrix resins and unique material forms that offer increased performance and improved processability over state-of-the-art structural composite materials. Two prepreg concepts will be developed and evaluated. The first will use either thermoplastic or thermoset polymer powders to impregnate fiber tows or woven preforms. The second will employ thermoplastics spun into fibers and intimately blended with the reinforcing fibers.

EFFECTS OF MATRIX AND INTERPHASE ON CARBON FIBER COMPOSITE COMPRESSION STRENGTH
NAS1-18883
1989 April - 1995 May

Project Engineer: Dr. Jeff Hinkley
Mail Stop 226
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-4259 FTS 928-4259

Principal Investigator: Dr. Willard D. Bascom
M.S. 304 EMRO
University of Utah
Salt Lake City, UT 84112
(801)-581-7422

Objective: Determine the material parameters that control composite compression strength, so that the fiber/matrix/interphase combination can be optimized. At least four commercial carbon fibers, five fiber coatings, and several epoxy matrix resins having different moduli will be studied. In each case, individual fiber failure modes will be characterized using in-situ microscopy and post-failure etching techniques. Next, propagation of damage in small bundles of fibers and in individual plies will be examined using specially-constructed specimens. Finally, the translation of these effects to strengths of laminate coupons and to multiaxial tension-compression behavior of tube structures will be determined.

CHARACTERIZING THE FRACTURE TOUGHNESS OF ADVANCED COMPOSITE STRUCTURES
NAS1-18833
1989 April - 1995 May

Project Engineer: Dr. John Crews
Mail Stop 188E
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3457 FTS 928-3457

Principal Investigator: Dr. John A. Nairn
M.S. 304 EMRO
University of Utah
Salt Lake City, UT 84112
(801) 581-3413

Objective: Develop fracture mechanics analyses for predicting matrix microcracks and microcrack-induced delaminations. Conduct tests to identify the parameters that govern microcracking and microcrack-induced delaminations. Then, develop strain energy release rate (G) analyses for observed damage initiation modes and growth modes. Finally, interpret composite stiffness degradation and fracture toughness in terms of critical strain energy release rates for damage initiation and growth.

THE MICROMECHANICS OF FATIGUE FAILURE IN WOVEN AND STITCHED COMPOSITES
NAS1-18840
1989 April - 1995 May

Project Engineer: Dr. Charles E. Harris
Mail Stop 188E
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3449 FTS 928-3449

Principal Investigator: Dr. Brian Cox
P.O. Box 1085
Rockwell International
1049 Camino Dos Rios
Thousand Oaks, CA 91360
(805)-373-4128

Objective: Develop experimental techniques to characterize the initiation and growth of fatigue damage. Determine the effect of damage on the internal stresses and the global composite stiffness. Based on damage characterization, develop micromechanical model for predicting fatigue behavior of new material architectures.

DAMAGE TOLERANCE OF COMPOSITE PLATES
NAS1-18778
1989 April - 1995 May

Project Engineer: Mr. C. C. Poe
Mail Stop 188E
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3467 FTS 928-3467

Principal Investigator: Dr. G. S. Springer
Department of Aeronautics and Astronautics
Stanford University
Stanford, CA 94305
(415)-723-4135

Objective: Develop an analysis to predict the complete damage state during and after low-velocity impact and to predict the residual properties. The analysis will be sufficiently general to account for the unique properties of thermoplastic matrix materials while applying to other matrices as well. A three-dimensional finite element model will be developed to calculate stresses, strains, and displacements in a composite during impact based on Hertzian contact forces. The model will define impactor position, velocity, and force as a function of time and will be general regarding material properties and composite layup. The model will predict fiber and matrix damage and trace delamination growth. The analysis will be verified through impact tests wherein the impact force and the extent of damage will be measured. Both destructive and nondestructive techniques will be used to determine the extent of damage.

ADVANCED FIBER PLACEMENT FUSELAGE TECHNOLOGY PROGRAM
NAS1-18887
1989 April - 1995 May

Project Engineer: Mr. W. T. Freeman
Mail Stop 241
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-2945 FTS 928-2945

Principal Investigator: R. L. Anderson
M.S. X11K4
Hercules Incorporated
P.O. 98
Magna, Utah 84044
(801)-251-2077

Objective: To develop breakthrough technology for cost effective fabrication of damage tolerant composite fuselage structures. A six-axis tow placement technique will be used to achieve low cost manufacturing of highly efficient complex structural forms. Major emphasis shall be on innovative manufacturing methods that may offer options for highly efficient primary aircraft structures. Six 3 x 4-ft flat isogrid panels will be fabricated using 8551-7/IM-7 damage tolerant material and subjected to a variety of static tests with and without damage. Following flat panel qualification in Phase 1, three full scale 5 x 6-ft curved panels will be fabricated and tested for concept verification.

INNOVATIVE FABRICATION PROCESSING OF ADVANCED COMPOSITE MATERIALS CONCEPTS FOR PRIMARY AIRCRAFT STRUCTURE
NAS1-18799

1989 May 9 - 1992 August 9

Project Engineer: Mr. Jerry. W. Deaton
Mail Stop 188B
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-30870 FTS 928-3087

Principal Investigator: Mr. S. P. Garbo
United Technologies
Sikorsky Aircraft Division
6900 Main Street
Stratford, Conn. 06601-1381
(203)-386-4576

Objective: Develop unique and innovative design concepts for complex fuselage structure that are amenable to the Therm-X pressure molding fabrication process. Concept drivers include innovative structural arrangement, improved structural efficiency, damage resistance, maintainability and repairability, and lower fabrication costs. Integrated design and Therm-X fabrication process to produce fuselage structure with frame/stringer intersections in a single cure operation.

MODELING AND DESIGN ANALYSIS METHODOLOGY FOR COMPOSITE PRIMARY STRUCTURE
NAS1-18754
1989 April - 1995 May

Project Engineer: Mr. O. F. Lopez
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3181 FTS 928-3181

Principal Investigator: Dr. L. W. Rehfield
Dept. of Mechanical Engineering
University of California
Davis, CA. 95616
(916)-752-0580

Objective: Develop and validate new structural models for aeroelastically tailored wings. Incorporate the non-classical deformation coupling modes of bending-transverse shear and extension-transverse shear into finite element structural analysis programs and account for section camber deformations in the analysis. Develop simple analytical models, useful for preliminary design and trade-off studies, that account for the essential physical behavior of the structure.

STUDY OF TAILORED COMPOSITE STRUCTURES OF ORDERED STAPLE THERMOPLASTIC MATERIALS
NAS1-18758
1989 April - 1995 May

Project Engineer: Ms. D. C. Jegley
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3185 FTS 928-3185

Principal Investigator: Dr. M. H. Santare
Dept. of Mechanical Engineering
University of Delaware
Newark, DE. 19716
(302)-451-2246

Objective: Develop and verify an analysis method to predict the response of curved beam structures that accounts for beams with various cross sections, microstructures, anisotropy and position-dependent material properties. Design curved beam test specimens made of ordered staple thermoplastic materials. Develop a methodology for fabrication of these test specimens that makes use of cost-effective manufacturing and sheet-forming technology.

ADVANCED TECHNOLOGY COMPOSITE AIRCRAFT STRUCTURES
NAS1-18889
1989 April - 1995 May

Project Engineer: Mr. W. T. Freeman
Mail Stop 241
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-2945 FTS 928-2945

Principal Investigator: Mr. P. J. Smith
Boeing Commercial Airplanes
M.S. 6N-21
P.O. Box 3707
Seattle, Wash. 98124
(206) 234-6733

Objective: To support NASA's goal to revitalize the nation's capacity for aeronautical innovation over the next decade by developing technology needed to apply composites to primary structures on commercial transport aircraft by the late 1990's. The technology shall provide a high level of technical confidence and demonstrate acceptable cost effectiveness for these specific objectives: (1) Develop basic technologies required to support cost effective damage tolerant pressurized fuselage structural designs and verify breakthrough technology results with mechanical tests. (2) Demonstrate advanced material placement processes and flexible automation for low cost assembly of pressurized transport fuselage structures. (3) Demonstrate the use of thermoplastic materials with advanced manufacturing techniques for fuselage clips, fittings, frames, and window belt reinforcements. (4) Develop the associated design, analysis and process technologies so that commercial application readiness and cost effectiveness can be realistically assessed. (5) Since the fuselage has the highest percentage of corrosion and fatigue problems on transport aircraft, composites will be evaluated for their potential to reduce repair and maintenance costs associated with airline life-cycle supportability. (6) Composite center fuselage elements will be developed because weight reductions at the airplane centerline are more effective in increasing payload, due to the offsetting dead-weight relief effects.

INNOVATIVE COMPOSITE AIRCRAFT PRIMARY STRUCTURES (ICAPS)
NAS1-18862
89 March 31 - 94 September 30

Project Engineer: Mr. Marvin B. Dow
Mail Stop 188B
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3090 FTS 928-3090

Principal Investigator: Mr. Raymond J. Palmer
McDonnell Douglas Corporation
Douglas Aircraft Company
3855 Lakewood Blvd.
Long Beach, CA 90846
(213) 593-7232

Objective: Develop and demonstrate innovative woven/stitched fiber preforms and resin matrix impregnation concepts for transport wing and fuselage structures. Demonstrate tow placement processes for transport fuselage structures. Demonstrate the use of thermoplastic materials with advanced manufacturing techniques for fighter aircraft fuselage structures.

NOVEL COMPOSITES FOR WING AND FUSELAGE APPLICATIONS
NAS1-18784
89 April 28 - 93 July 31

Project Engineer: Mr. H. Benson Dexter
Mail Stop 188B
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3094 FTS 928-3094

Principal Investigator: Mr. James Suarez
Grumman Aerospace Corporation
Aircraft Systems Division
South Oyster Bay Road
Bethpage, NY 11714-3582
(516) 346-3941

Objective: Integrate innovative design concepts with cost-effective fabrication processes to achieve damage tolerant structures that can perform at a design ultimate strain level of at least 6000 micro in./in. Integral structures will be fabricated using weaving and knitting/stitching concepts. Resin transfer molding will be used for low cost resin application and consolidation.

INNOVATIVE COMPOSITE FUSELAGE STRUCTURES
NAS1-18842
1989 April - 1995 May

Project Engineer: Mr. M. Rouse
Mail Stop 190
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3182 FTS 928-3182

Principal Investigator: Mr. R. B. Deo
M.S. 3853/82
Northrop Corporation
1 Northrop Ave
Hawthorne, CA. 90250-3277
(213)-332-2134

Objective: Develop innovative concepts for fighter aircraft fuselage structures that will improve structural efficiency while reducing manufacturing costs. Analysis methods and structural mechanics methodologies appropriate for the new structural concepts will also be developed and validated through tests of elements and components. Scaling laws that account for scale-up effects to predict the overall failure of built-up fuselage structure based on subscale tests will be developed. At least five design concepts will be selected and a detailed assessment made of their potential for meeting the program goals. Various matrix resins and material forms will be considered including toughened thermosets, thermoplastics, bismaleimides, polyimides, and polycrystalline materials in unidirectional and woven prepregs and woven, stitched, or braided preforms.Analysis techniques will be developed in three major areas: (1) structural details, (2) structural stability, and (3) scaling laws. Structural details to be considered include corner radii, lay-up discontinuities such as ply drop-offs and stiffener terminations, and stresses at cut-outs. Stability related analyses will be developed dealing with general instability, local-global buckling interactions, stiffener crippling, and stiffener/skin separation. Scale-up effects will be investigated through a building-block approach. Each structural detail will be analyzed i independently; then, analyses will be developed to predict probable failure sequences in large-scale built-up structure that account for load redistribution subsequent to first element failure.

AIR FORCE OFFICE OF SCIENTIFIC RESEARCH

IN-HOUSE

NONE

GRANTS AND CONTRACTS

DIRECT OBSERVATION OF CRACKING AND THE DAMAGE MECHANICS OF CERAMICS AND CERAMIC COMPOSITES
AFOSR-87-0307
01 June 87 - 31 May 90

Principal Investigator: Dr Peter W R Beaumont
Engineering Department
Cambridge University
Trumpington Street, Cambridge CB2 1PZ
(01144) 223-332600

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To directly observe and analyze microcracking and spalling in ceramic materials. To study inherent toughening methods, such as plasticity in oxide ceramics at high temperature, constrained plasticity, soft cobalt matrices, and localized transformation toughening.

INTERFACIAL STUDIES OF WHISKER REINFORCED CERAMIC MATRIX COMPOSITES
FQ8671-8900267
1 May 88 - 30 April 90

Principal Investigator: Dr John Brennan
United Technologies Research Center
East Hartford, CT 06108
(203) 727-7220

Program Manager: Dr Liselotte J Schioler
AFOSR/NE
Bolling AFB DC 20332-6448
(202) 767-4933

Objective: To study the microstructural and microchemical properties of interfaces of SiC and Si_3N_4 whisker reinforced glass and glass-ceramic matrix composites.

TEMPERATURE PERFORMANCE OF CERAMIC AND GLASS MATRIX COMPOSITES
AFOSR-87-0383
15 July 87 - 14 October 91

Principal Investigator: Professor Tsu-Wei Chou
University of Delaware
Center for Composite Materials
Newark, DE 19716
(302) 451-2904

Program Manager: Dr Liselotte J Schioler
AFOSR/NE
Bolling AFB DC 20332-6448
(202) 767-4933

Objective: To provide a fundamental understanding of the high-temperature mechanical properties, environmental effects and failure mechanisms of glass and ceramic matrix composites through experimental characterization and theoretical modeling, and to establish high-temperature mechanical testing and characterization methods for glass and ceramic matrix composites.

3D ANALYSIS AND VERIFICATION OF FRACTURE GROWTH MECHANISMS IN FIBER-REINFORCED CERAMIC COMPOSITES
AFOSR-89-0005
01 September 88 - 31 August 91

Principal Investigator: Professor Michael P Cleary
Department of Mechanical Engineering
Massachusetts Institute of Technology
Cambridge, MA 02139
(617) 253-2308

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To model the fracture mechanisms in mechanical systems representative of existing and proposed ceramic composites. Emphasis is placed on the roles of the fibers and the interface in generating, arresting, or retarding the growth of fractures.

HETEROGENEOUS CHARACTERIZATION OF COMPOSITE MATERIALS WITH PROGRESSIVE DAMAGE
AFOSR-88-0124
01 February 88 - 31 January 91

Principal Investigator: Dr Isaac M. Daniel
Department of Civil Engineering
Northwestern University
Evanston, IL 60201
(312) 491-5649

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To develop constitutive and failure models for composite materials based on observed damage mechanisms and damage development. The study will include organic matrix composite materials such as graphite/epoxy, as well as high-temperature composites, such as ceramic matrix/ceramic fiber composites.

MICROMECHANICAL PREDICTION OF TENSILE DAMAGE FOR CERAMIC MATRIX
AFOSR-90-00895
15 August 90 - 14 August 93

Principal Investigators: Dr Feridun Delale
Dr Been-Ming Liaw
Department of Mechanical Engineering
The City College of
The City University of New York
New York, NY 10036
(212) 690-4252

Program Manager: Dr Liselotte J. Schioler
AFOSR/NE
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To model the failure mechanisms of microcracking at the fiber/matrix level in a ceramic-fiber/ceramic-matrix composite material subjected to thermo-mechanical loading.

DYNAMICS AND AEROELASTICITY OF COMPOSITE STRUCTURES
F49620-86-C-0066
01 July 86 - 30 September 90

Principal Investigator: Dr John Dugundji
Department of Aeronautics & Astronautics
Massachusetts Institute of Technology
Cambridge, MA 02139
(617) 253-3758

Program Manager: Dr Spencer T. Wu
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-6962

Objective: To pursue combined experimental and theoretical investigations of aeroelastic tailoring effects on flutter and divergence of aircraft wings.

DEFORMATION AND DAMAGE MECHANISMS IN HIGH TEMPERATURE COMPOSITES WITH DUCTILE MATRICES
AFOSR-88-0150
01 March 88 - 28 February 91

Principal Investigator: Dr George J Dvorak
Department of Civil Engineering
Rensselaer Polytechnic Institute
Troy, NY 12181
(518) 266-6943

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To develop a basis of understanding of the thermo-mechanical behavior of fibrous composites with ductile matrices and ductile or elasto-brittle fibers, and of the damage mechanisms activated by combined mechanical and thermal loading, both cyclic and monotonic.

COATINGS FOR FIBERS IN HIGH-TEMPERATURE, HIGH-PERFORMANCE COMPOSITES
DARPA Program
1 July 89 - 30 June 92

Principal Investigator: Professor Anthony Evans
U California, Santa Barbara
Santa Barbara, CA 93106
(805) 961-4634

Program Manager: Dr Liselotte J Schioler
AFOSR/NE
Bolling AFB DC 20332-6448
(202) 767-4933

Objective: To identify viable fiber coatings for composite systems, based on fundamental thermochemical and thermomechanical considerations, coupled with measurements made on experimental models and actual composite systems.

FAILURE CRITERIA IN LAMINATES BASED ON A 3-D MICROMECHANICS CONSIDERATION
AFOSR-90-073
15 June 90 - 14 June 92

Principal Investigator: Dr E S Folias
Department of Civil Engineering
The University of Utah
Salt Lake City, UT 84112
(801) 581-6931

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To use analytically determined three-dimensional stress field derived for laminated composites to establish failure criteria based on micromechanics considerations.

COMPOSITE MATERIALS INTERFACE MECHANICS
AFOSR-MIPR-89-0022 (Co-funded with ONR)
01 March 89 - 28 February 91

Principal Investigator: Dr Zvi Hashin
Department of Mechanical Engineering and Applied Mechanics
University of Pennsylvania
Philadelphia, PA 19104-3246
(215) 898-8504

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To assess the effects of realistic interface conditions due to elastic, inelastic, and damaged interphase on the thermoelastic properties and failure of composite materials.

CRAZING IN POLYMERIC AND COMPOSITE SYSTEMS
AFOSR-87-0143
01 April 87 - 31 March 90

Principal Investigator: Dr C C Hsiao
Deparment of Aerospace Engineering and Mechanics
University of Minnesota
Minneapolis, MN 55455
(612) 625-7363

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To develop time-dependent theories for the crazing behavior of polymeric and structural composite systems by understanding the microstructural behavior of the materials during crazing.

DEVELOPMENT OF CAPABILITY FOR CHARACTERIZATION OF CERAMIC/CERAMIC COMPOSITES
F49620-89-C-0016
01 November 88 - 31 October 90

Principal Investigator: Dr Shaik Jeelani
Tuskegee Institute
Tuskegee, AL 36088
(205) 727-8970

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-4987

Objective: To devise testing methods for directly observing and characterizing damage mechanisms in fiber-reinforced ceramic composites.

HIGH TEMPERATURE HETEROGENEOUS MATERIALS
AFOSR-90-0237
01 December 89 - 30 November 92

Principal Investigator: Professor Leon M Keer
Department of Civil Engineering
Northwestern University
Evanston, IL 60208
(708) 491-4046

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To study the high-temperature behavior of ceramic fiber-reinforced, ceramic matrix composites. The types, severity, and growth of damage mechanisms under various loading conditions will be experimentally established and analytically described.

MECHANICS OF FAILURE OF HIGH TEMPERATURE METAL MATRIX COMPOSITES
AFOSR-89-424
15 March 90 - 14 March 93

Principal Investigator: Professor D A Kouris
Department of Mechanical and Aerospace Engineering
Arizona State University
Tempe, AZ 85287
(602) 965-4977

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To establish the damage mechanisms mainly responsible for the degradation of metal matrix composite materials under high-temperature fatigue conditions and to describe those mechanisms analytically so that predictions of the expected behavior can be made. Experimental work (to be carried out by Rockwell International Science Center) will establish physical damage mechanisms by direct observation and will also verify analytical damage-growth rates predictions.

THE MECHANICS OF PROGRESSIVE CRACKING IN CERAMIC MATRIX COMPOSITES AND LAMINATES
AFOSR-88-0104
01 February 88 - 31 January 91

Principal Investigator: Dr Norman Laws
Department of Mechanical Engineering
University of Pittsburgh
Pittsburgh, PA 15261
(412) 624-9793

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To study damage processes in continuous fiber-reinforced ceramic matrix composites (CMC), and, in particular, the degradation (or improvement) of thermo-mechanical properties when the composites have been damaged by matrix cracking, fiber debonding, and ultimately fiber pull-out.

HIGH PERFORMANCE LAMINATED COMPOSITES
AFOSR-90-0132
01 January 90 - 31 December 92

Principal Investigator: Professor F A Leckie
Department of Mechanical Engineering
University of California
Santa Barbara, CA 93106
(805) 961-2652

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To establish the mechanics framework which will allow the analysis and interpretation of the mechanical behavior of laminated systems consisting of thin alternating layers of brittle and ductile materials.

MICROMECHANICS OF INTERFACE IN HIGH-TEMPERATURE COMPOSITES
AFOSR-89-0269
01 February 89 - 31 January 92

Principal Investigator: Professor Toshio Mura
Department of Civil Engineering
Northwestern University
Evanston, IL 60208
(312) 491-4003

Program Managers: Lt Colonel George K Haritos & Dr Liselotte J Schioler

Lt Colonel George K Haritos	Dr Liselotte J Schioler
AFOSR/NA	AFOSR/NE
Bolling AFB DC 20332-6448	Bolling AFB DC 20332-6448
(202) 767-0463	(202) 767-4933

Objective: To establish the microstructural variables which promote toughness in brittle matrix composites by means of analytical and experimental perspectives, and to construct the mechanics/material sciences based model for predicting the behavior of such materials.

OPTIMUM AEROELASTIC CHARACTERISTICS FOR COMPOSITE SUPER-MANEUVERABLE AIRCRAFT
AFOSR-89-0055
01 October 88 - 30 September 91

Principal Investigator: Dr Gabriel Oyibo
Department of Mechanical & Aerospace Engineering
Polytechnic University
Farmingdale, NY 11735
(516) 454-5120

Program Manager: Dr Spencer T Wu
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-6962

Objective: To identify, characterize, and model the effects of constrained warping on the dynamics and aeroelastic stability of aircraft composite wings.

FINITE ELEMENT ANALYSIS OF COMPOSITE SHELLS
AFOSR-PD-88-0010
01 April 89 - 30 September 90

Principal Investigator: Dr Anthony Palazotto
Air Force Institute of Technology
Wright-Patterson AFB OH 45433-6583
(513) 255-2998, AUTOVON 785-2998

Program Manager: Dr Spencer T Wu
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-6962

Objective: A general nonlinear shell theory has been developed to analyze the static and dynamic characteristics of composite shells. A finite element program is being developed. Pertubation and boundary integral techniques are also being used for baseline and comparison purposes.

MESOMECHANICAL MODEL FOR FIBRE COMPOSITES: THE ROLE OF THE INTERFACE
AFOSR-89-0365
01 June 89 - 31 May 92

Principal Investigator: Professor M R Piggott
University of Toronto
Ontario, Canada M5S 1A4
(416) 978-4745

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To establish the relationship between the interface/interphase parameters and the composite properties. Particular attention is paid to the role of interphase failure.

EVOLUTION MECHANICS
AFOSR-89-0216
01 December 88 - 30 November 91

Principal Investigator: Professor K L Reifsnider
Virginia Polytechnic Institute & State University
Blacksburg, VA 24061
(703) 961-5316

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448

Objective: To develop methods for predicting the long-term behavior of composite materials, especially their remaining strength and life after periods of service which includes exposure to time-variable thermomechanical and chemical loading.

EIGENSENSITIVITY ANALYSIS OF COMPOSITE LAMINATES: EFFECT OF MICROSTRUCTURE
F49620-89-C-0003
01 November 88 - 31 October 90

Principal Investigator: Dr Robert Reiss
Howard University
Washington DC 20059
(202) 636-6608

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To assess the sensitivity of composite laminates' natural (for elastic models) and complex (for viscoelastic models) frequencies to small changes in the properties of their constituents.

MICROMECHANICAL ANALYSIS OF CERAMIC MATRIX COMPOSITES
F49620-88-C-0069
01 April 88 - 31 March 90

Principal Investigator: Dr B Walter Rosen
Materials Sciences Corporation
930 Havest Drive
Union Meeting Corporate Center
Blue Bell, PA 19422
(215) 542-8400

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To develop a material model of a unidirectional composite which accounts for residual stresses, matrix porosity, interphases, cracks perpendicular to fibers, cracks parallel to fibers, interface debonding, fiber fracture, and in general, the accumulation and growth of various types of damage.

DEFORMATION AND FRACTURE OF FIBER-REINFORCED CERAMIC COMPOSITES
AFOSR-87-0257
01 October 89 - 30 September 90

Principal Investigator: Dr Richard A Schapery
Department of Civil Engineering
Texas A&M University
College Station, TX 77843-3124
(409) 845-2449

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To study the thermo-mechanical fatigue behavior of ceramic fiber-reinforced, ceramic-matrix composites. The types, severity, and growth mechanisms of fatigue damage in the composites under various loading conditions will be experimentally characterized and analytically modelled.

DAMAGE ACCUMULATION IN ADVANCED METAL-MATRIX COMPOSITES UNDER THERMAL CYCLING
AFOSR-890059
15 October 88 - 14 October 91

Principal Investigator: Professor M Taya
University of Washington
Seattle, WA 98195
(206) 545-2850

Program Manager: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To characterize the mechanisms of the damage accumulation process in metal-matrix composites subjected to creep and/or thermal cycling, including the nucleation and growth of interface damage.

ANISOTROPIC DAMAGE MECHANICS MODELLING IN METAL MATRIX COMPOSITES
AFOSR-90-0227
01 April 90 - 31 March 93

Principal Investigator: Professor G. Z. Voyiadjis
Department of Civil Engineering
Louisiana State University
Baton Rouge, LA 70803
(504) 388-8668

Program Manger: Lt Colonel George K Haritos
AFOSR/NA
Bolling AFB DC 20332-6448
(202) 767-0463

Objective: To formulate a constitutive model for ductile fracture and finite strains in metal matrix composites, using the anisotropic theory of continuum damage mechanics. A damage tensor derived with respect to the unstressed damaged state will be utilized.

A COMPREHENSIVE STUDY OF MATRIX FRACTURE MECHANISMS IN FIBER-REINFORCED MATRIX COMPOSITES
AFOSR-90-0172
15 February 90 - 14 February 93

Principal Investigator: Dr Albert S D Wang
Department of Mechanical Engineering and Mechanics
Drexel University
Philadelphia, PA 19104
(215) 895-2297

Program Manager:	Lt Colonel George K Haritos AFOSR/NA Bolling AFB DC 20332-6448 (202) 767-0463
Objective:	To establish both a fabrication and a material characterization capability for a class of high-temperature ceramic matrix composites as an integrated effort.

ARMY PROGRAM INPUT

Fifteenth Annual Mechanics of Composites Review

Dayton, Ohio

24 - 25 October 1990

submitted by
Dr. Bruce P. Burns
U.S. Army Ballistic Research Laboratory
ATTN: SLCBR-IB-M
Aberdeen Proving Ground, MD 21005-5066

U. S. ARMY
ARMY RESEARCH OFFICE

CONTRACTS

TITLE: Advanced Mechanical Design of High Performance Articulating Robotic Systems
RESPONSIBLE INDIVIDUAL G. L. Anderson
Army Research Office
P. O. Box 12111
Research Triangle Park, NC 27709-2211
(919) 549-0641
PRINCIPAL INVESTIGATOR: M. V. Gandhi and B. S. Thompson
Dept. of Mechanical Engineering
Michigan State University
Ease Lansing MI 48824-1226

OBJECTIVE: Develop a new generation of high performance light weight robot arms fabricated in advanced composite materials.

TITLE: Stability of Elastically Tailored Rotor Systems
RESPONSIBLE INDIVIDUAL G. L. Anderson
Army Research Office
P. O. Box 12111
Research Triangle Park, NC 27709-2211
(919) 549-0641
PRINCIPAL INVESTIGATOR: D. Hodges and L. Rehfield
School of Aerospace Engineering
Georgia Institute of Technology
Atlanta, GA 30332
OBJECTIVE: Develop mathematical modeling and analysis procedures to determine the aeroelastic stability characteristics of bearingless helicopter rotors on elastic supports in axial flow and tilt rotor aircraft with elastic wings in axial flight in the helicopter mode and in the airplane mode. The rotor systems are composed of or contain significant structural components fabricated from composite materials.

TITLE: Damage Resistance in Rotorcraft Structures
RESPONSIBLE INDIVIDUAL G. L. Anderson
Army Research Office
P. O. Box 12111
Research Triangle Park, NC 27709-2211
(919) 549-0641
PRINCIPAL INVESTIGATOR: E. A. Armanios
School of Aerospace Engineering
Georgia Institute of Technology
Atlanta, GA 30332

OBJECTIVE: Explore the benefits of tailoring microstructure, i.e., ply stacking se-quence, fiber orientation, and a blend of material plies, to contain and resist damage in rotor systems and airframe structural components. The analysis will be developed for a generic damaged ply model that includes matrix micro-cracking, delamination and fiber fracture, and their interaction.

TITLE: Optimization of Composite Drive Shafts
RESPONSIBLE INDIVIDUAL G. L. Anderson
Army Research Office
P. O. Box 12111
Research Triangle Park, NC 27709-2211
(919) 549-0641
PRINCIPAL INVESTIGATOR: M. Darlow and O. A. Bachau
Dept. of Mechanical Engineering
Renssalaer Polytechnic Institute
Troy NY 12180-3590

OBJECTIVE: Develop a computerized design process for designing composite drive shafts for rotorcraft that can operate at super-critical rotational speeds. Develop algorithms to optimize shaft systems based on geometric envelope, torsional strength and elastic stability (buckling), torsional and lateral vibrations, and weight.

TITLE: Analysis and Design of Composite Fuselage Frames
RESPONSIBLE INDIVIDUAL G. L. Anderson
Army Research Office
P. O. Box 12111
Research Triangle Park, NC 27709-2211
(919) 549-0641
PRINCIPAL INVESTIGATOR: O. A. Bachau
Dept. of Mechanical Engineering
Renssalaer Polytechnic Institute
Troy NY 12180-3590

OBJECTIVE: Develop a model that will allow the accurate analysis and design of helicop-ter fuselage frame components using composite materials. The features to be included are strong curvature (height to radius of curvature ratio of the order of 1 to 3), major secondary stresses (crushing and curling) due to this curvature, sharp changes in gage thickness, and material anisotropy effects including continuously varying directions of principal axes of orthotropy.

TITLE: Hygrothermal Effects on the Elastic Properties of Tailored Composite Blades
RESPONSIBLE INDIVIDUAL: G. L. Anderson
Army Research Office
P. O. Box 12111

Rensselaer Polytechnic Institute
Troy, NY 12180-3590

OBJECTIVE: Develop analytical models for predicting changes in stiffness and coupling during hygrothermal conditioning. Perform experiments to measure such effects.

TITLE: Formal Optimization Procedures for Composite Blades
RESPONSIBLE INDIVIDUAL: G. L. Anderson
Army Research Office
P. O. Box 12111
Research Triangle Park, NC 27709-2211
(919) 549-0641
PRINCIPAL INVESTIGATOR: O. A. Bauchau
Dept. of Mechanical Engineering
Rensselaer Polytechnic Institute
Troy, NY 12180-3590

OBJECTIVE: For the full application of composite materials to rotor blades, develop design and optimization tools that allow for the imposition of multiple constraints.

TITLE: Advanced Composite Laminates for Rotorcraft
RESPONSIBLE INDIVIDUAL: G. L. Anderson
Army Research Office
P. O. Box 12111
Research Triangle Park, NC 27709-2211
(919) 549-0641
PRINCIPAL INVESTIGATOR: R. J. Diefendorff, O. A. Bauchau and S. J. Winkler

Dept. of Mechanical Engineering
Rensselaer Polytechnic Institute
Troy, NY 12180-3590

OBJECTIVE: Analytical modeling, fabrication and testing research will be undertaken to develop new two and three dimensional composite concepts that promise advanced elastic tailoring, improved load transfer and/or reduced fabrication costs. Analysis method-ology will be developed that is capable of predicting the elastic characteristics of laminates with "bend" and "splay" intralaminate fiber concepts.

TITLE: Finite Element Modelling of Composite Rotor Blades
RESPONSIBLE INDIVIDUAL: G. L. Anderson
Army Research Office
P. O. Box 12111
Research Triangle Park, NC 27709-2211
(919) 549-0641
PRINCIPAL INVESTIGATOR: S. Lee and A. Vizzini
Dept. of Aerospace Engineering
University of Maryland
College Park, MD 20742

OBJECTIVE: Develop a beam finite element formulation of the combined dynamic bending, torsional, and extensional behavior of composite rotor blades taking into account the warping effect of blades undergoing large deflection or finite rotation. This new approach models thin walled composite blades with complicated cross sections, tapers, and arbitrary planforms. The warping effect is incorporated by assuming warping displace-ments superimposed over cross sections normal

TITLE: Manufacturing Science, Reliability and Maintainability Technology RESPONSIBLE INDIVIDUAL: A. Crowson
Army Research Office
P. O. Box 12111
Research Triangle Park, NC 27709-2211
(919) 549-0641
PRINCIPAL INVESTIGATORS: T. W. Chou and R. L. McCullough
Center For Composite Materials
University of Delaware
Newark DE 19716

OBJECTIVE: This University Research Initiative Program consists of the following elements : cure characterization and monitoring, on-line intelligent non destructive evaluation for in-process control of manufacturing, process simulation, computer aided manufacturing by filament winding, structural property relationships, mechanics of thick section composite laminates, structure performance and durability and integrated enginee-ring for durable structures.

U. S. ARMY LABORATORY COMMAND
MATERIALS TECHNOLOGY LABORATORY

TITLE: Composite Hull for Light Infantry Fighting Vehicle
PROJECT ENGINEER: W. Haskell
U.S. Army Laboratory Command
Materials Technology Laboratory
Watertown, MA 02172-0001
(617) 923-5172
PRINCIPAL INVESTIGATOR: E. Weerth
FMC Corp.
San Jose, CA OBJECTIVE: Demonstrate the application of thick composites technology to armored vehicles for the purpose of weight reduction. Payoffs in the form of reduced weight over aluminum for equal ballistic protection, reduced spall, elimination of corrosion, signature reduction, reduced life cycle costs and logistic improvements related to easier transportability and lowered fuel consumption, are being demonstrated.

TITLE: Design Analysis of Composite Test Specimens
PROJECT ENGINEER: D. W. Oplinger
U.S. Army Laboratory Command
Materials Technology Laboratory
Watertown, MA 02172-0001
(617) 923-5303
PRINCIPAL INVESTIGATOR: S. Chaterjee
Materials Sciences Corp.
Gwynedd Plaza II
Bethlehem Pike
Spring House PA
(215) 542-8400

OBJECTIVE:The objective is to evaluate current specimen designs for mechanical-property test specimens for composites and to develop design improvements. Effort includes combined stress analysis effort to evaluate specimen designs of interest, and experimental effort to provide confirmatory data, both for evaluation of current specimens and assessment of suggested

TITLE: Failure and Degradation of Adhesive Joints
RESPONSIBLE INDIVIDUAL: A. Johnson
U.S. Army Laboratory Command
Materials Technology Laboratory
Watertown, MA 02172-0001
(617) 923-5427
PRINCIPAL INVESTIGATOR: D. Oplinger
U.S. Army Laboratory Command
Materials Technology Laboratory
Watertown, MA 02172-0001
(617) 923-5259

OBJECTIVE:The objective is to provide the Army with improved design, life prediction and reliability methodology for adhesive joints. The effort includes analytical and experimental efforts aimed at investigating:(1) development of improved analytical methods for fracture of adhesive joints, including adhesive bond failures as well as cohesive failures in composite adherends;(2) application of moire interferometry to evaluating pre-cracked bending specimens for adhesive testing;(3) provision of an up-to-date assess-ment of joint stress analysis and design methodology; (4) development of methodology for 3-D modelling of joints;(5) investiga-tion of environmental degradation effects;(6) develop-ment of improved finite element approaches for modelling adhesive joints.

TITLE: Computational Mechanics of Thick Composites
RESPONSIBLE INDIVIDUAL: A. Johnson
U.S. Army Laboratory Command
Materials Technology Laboratory
Watertown, MA 02172-0001
(617) 923-5427
PRINCIPAL INVESTIGATOR: A. Tessler
U.S. Army Laboratory Command
Materials Technology Laboratory
Watertown, MA 02172-0001
(617) 923-5356

OBJECTIVE: The program objective is to develop analytic methods for predicting mechanical response and failure of advanced thick composite structures which may involve stress concentrators such as cutouts, fasteners, delaminations and defects. The analytic development is facilitated by a comprehensive experimental verification involving modal analysis and moire methods. The program approach encompasses: (1) development of an effective and yet simple higher-order laminated composite shell theory which would be amenable to finite element modeling to simulate linear and nonlinear dynamic response; (2) development of reliable and efficient finite element models for the analysis of composite shell structures and adhesively bonded composite joints; (3) experimental validation of the analytic and computational models via modal analysis and moire strain methods. The technology will improve the design methods for Army's thick composite structures such as those employed in helicopter rotor blades, tank hulls and turrets, light-weight howitzers, kinetic energy projectiles, and a whole range of other applications.

TITLE: Dynamics of Structures
RESPONSIBLE INDIVIDUAL: R. Shufford
U.S. Army Materials Technology Laboratory
ATTN: SLCMT-MEC
Watertown, MA 02172
(617) 923-5514

OBJECTIVE: The objective of this program is the development of modal analysis as a predictive technique for detection of damage in composite structures, such as foam core sandwich panels. This is done by characterizing the structure in terms of its modal parameters. Changes in the damping, natural frequencies, as well as the mode shapes are investigated as a function of damage in composite structures.

TITLE: Effect of Fabrication Variables on Composite Structures
RESPONSIBLE INDIVIDUAL: R. Shufford
U.S. Army Materials Technology Laboratory
ATTN: SLCMT-MEC
Watertown, MA 02172
(617) 923-5572
PRINCIPAL INVESTIGATORS: G.E. Foley, S. Ghiorse, M.E. Roylance
U.S. Army Materials Technology Laboratory
ATTN: SLCMT-MEC
Watertown, MA 02172
(617) 923-5514

OBJECTIVE: The objective of this program is to determine the advantages and/or disadvantages of different manufacturing techniques, such as braiding, filament winding, and hand lay-up on the mechanical properties, as well as the environmental durability of these composites.

TITLE: Automated Evaluation of Composite Materials
RESPONSIBLE INDIVIDUAL: G.L. Hagnauer
U.S. Army Materials Technology Laboratory
ATTN: SLCMT-EMP
Watertown, MA 02172-0001
PRINCIPAL INVESTIGATORS: G.L. Hagnauer and S.G.W. Dunn
Polymer Research Branch
U.S. Army Materials Technology Laboratory
(617) 923-5121

OBJECTIVES: The objectives of this project are to increase laboratory productivity and improve the quality of test information needed to evaluate fiber-reinforced polymeric matrix composite materials and guide their specification, design and manufacture. Laboratory robotics and artificial intelligence (AI) technologies are being developed to meet requirements for handling and testing large numbers of specimens under a wide range of conditions and to increase the efficiency and reduce labor costs involved in evaluating composite materials. To control automation and deal with the large amounts of information generated by automated testing, advanced computer and AI technologies (e.g., expert systems, image analysis and machine learning) are being implemented. AI techniques will be employed to advise and plan tests, control robots and automated test equipment, and interpret and preserve information in a living database and as reports with fully traceable documenta-tion. Currently, the technology is being implement-ed in research on the durability evaluation and life prediction of composites.

TITLE: Dynamic characterization of Advanced Materials
RESPONSIBLE INDIVIDUAL: J. Nunes

PRINCIPAL INVESTIGATORS: W. Crenshaw
U.S. Army Materials Technology Laboratory
ATTN: SLCMT-MRM-MTG
Watertown, MA 02172-0001
(617) 923 5203

OBJECTIVE: Design and evaluate instrumentation systems and experimental designs to measure load response of advanced composites and homogeneous materials to low speed impact loading. Conduct standard material evaluations to determine residual strength oaf the material after impact. Determine the accuracy if present constitutive models in predicting the dynamic behavior of composites and homogeneous materials during low speed impact.

TITLE: Ultrasonic Digital Signal Processing(DSP)
RESPONSIBLE INDIVIDUAL: A. Broz
U.S. Army Materials Technology Laboratory
ATTN: SLCMT-MRM
Watertown, MA 02172-0001
(617) 923 5285
PRINCIPAL INVESTIGATORS: B. Taber
U.S. Army Materials Technology Laboratory
ATTN: SLCMT-MRM
Watertown, MA 02172-0001
(617) 923 5443

OBJECTIVE: Digital Signal Processing(DSP) is a computer based technique for enhancement of digital signals related to image processing. DSP techniques are being developed that enhance ultrasonic inspection capabilities in conjunction with NDE of composites and other applications. For example, DSP can provide additional information about individual plies in thin-lamina composites. It has also been used for the evaluation of thin bondlines in adhesive bonds. NDE efforts at MTL will emphasize the use of DSP in conjunctions with ultrasonic B scans which have been used for void location, volume fraction determination and imaging of the layered structure of laminates.

U. S. ARMY LABORATORY COMMAND
BALLISTIC RESEARCH LABORATORY

TITLE: Lightweight Structures for Interior Ballistics

PRINCIPAL INVESTIGATOR: W.H. Drysdale
AMC LABCOM
Ballistic Research Laboratory
Aberdeen Proving Ground
MD 210055066
(301) 278-6123

OBJECTIVE: Composite materials represents a portion of this effort. The objective of this project is to develop failure criteria, architecture transition technology, and optimum design technology for thick ballistic structures. Rate of loading and layup transition studies are being addressed at BRL. A special, high-rate, propellant driven test apparatus is under development to generate uniaxial or triaxial stress states at strain rates of up to 200 per second. Three dimensional failure criteria and other constitutive effects are being studied and hypothesized by Lawrence Livermore National Lab(LLNL). They are also sponsoring studies at the University of Utah and Pennsylvania State University. Experimental activities to develop failure data are being conducted at both the LLNL and the University of Utah. Additional failure criteria work and extensions to optimal notions for relatively simple structures and layup.

TITLE: Determination of Mechanical Material Properties for Filament Wound Structures
RESPONSIBLE INDIVIDUAL: Dr. Larry C. Mixon
Army Missile Command
PRINCIPAL INVESTIGATOR: Terry L. Vandiver
Army Missile Command
(205) 876-1015

OBJECTIVE: The objective of this task is to develop test standards for the determination of mechanical material properties for filament wound composite structures. The initial task is to develop uniaxial material properties. Future plans include biaxial and triaxial material property determination. This effort is being performed by the Joint-Army-Navy-NASA-Air Force (JANNAF) Composite Motor Case Subcommittee through a round robin test effort. This task is coordinated with MIL-HDBK-17, ASTM, National Bureau of Standards, and DoD CMPS Composites Technology Program.

TITLE: Composite Materials Evaluation for Filament Winding
RESPONSIBLE INDIVIDUAL: Lawrence W. Howard
Army Missile Command
PRINCIPAL INVESTIGATOR: Terry L. Vandiver
Army Missile Command
(205) 876-1015

OBJECTIVE: The object of this task is to evaluate new fibers for filament winding. Delivered strengths are determined via strand tests and 3-inch diameter filament wound pressure vessels with different stress ratios. The experimental data is used in the design of composite rocket motor cases, launchers, pressure vessels and other filament wound structures.

TITLE: Composite Wing Design and Fabrication
RESPONSIBLE INDIVIDUAL: Lawrence W. Howard
Army Missile Command
PRINCIPAL INVESTIGATORS: J. Frank Wlodarski
Terry L. Vandiver
Army Missile Command
(205) 876-0398

OBJECTIVE: The objective of this task is to design and fabricate an all composite wing with an elliptical planform. The materials used are s-glass cloth and uni-directional tape. These materials were selected because of their strength, stiffness and low radar cross-section. The method of fabrication is hand layup in a clamshell mold made of composite tooling. The wings are tested to determine what structural properties are achieved with this method of manufacture and if they are accurately predicted in the design.

U.S. ARMY AVIATIONS SYSTEMS COMMAND
FT. EUSTIS

TITLE: Damage Tolerance Testing of the ACAP Roof

RESPONSIBLE INDIVIDUAL: F. Swats
U.S. Army ARTA (AVSCOM)
Aviation Applied Technology
Directorate
SAVRT-TY-ATS
Ft. Eustis, VA 23604-5577
(804) 878-2975
PRINCIPAL INVESTIGATOR: B. Spigel

OBJECTIVE: A forward roof subcomponent from the Bell Advanced Composite Airframe Program (ACAP) helicopter will be tested to verify the damage tolerance design criteria developed under contract by Bell Helicopter Textron, Inc. (Final Report: USAAVSCOM TR-87-D-3A, B, C). The roof will be subjected to an anticipated ACAP load spectrum, and manufacturing defects and in-service damage will be monitored by both laboratory and field nondestructive evaluation methods to determine the extent of damage growth.

TITLE: Ballistic Survivability of Generic Composite Main Rotor Hub Flexbeams
RESPONSIBLE INDIVIDUAL: F. Swats
U.S. Army ARTA (AVSCOM)
Aviation Applied Technology
Directorate
SAVRT-TY-ATS
Ft. Eustis, VA 23604-5577
(804) 878-2975
PRINCIPAL INVESTIGATORS: E. Robeson and K. Sisitka

OBJECTIVE: The goal of this effort is to quantify the ballistic survivability of typical composite main rotor hub flexbeams. Two different flexbeam designs will be impacted with various ballistic threats. One design will be tested under simulated centrifugal load while the other will be fatigue tested following ballistic impact in a no load condition. Fatigue testing of the first design will be considered after a damage assessment is made.

TITLE: Finite Element Correlation of the Advanced Composite Airframe Program (ACAP) Dynamic Models
RESPONSIBLE INDIVIDUAL: E. Austin
U.S. Army ARTA (AVSCOM)
Aviation Applied Technology Directorate
SAVRT-TY-ATS
Ft. Eustis, VA 23604-5577
(804) 878-3822
PRINCIPAL INVESTIGATORS: N. Calapodas and D. Kinney
U.S. Army ARTA (AVSCOM)
Aviation Applied Technology Directorate
SAVRT-TY-ATS
Ft. Eustis, VA 23604-5577
(804) 878-3303

OBJECTIVE: A joint program among Army/NASA/Contractor is planned to conduct detail correlation of the Finite Element (FE) dynamic models of both ACAP airframes. AATD will perform all shake testing and the contractors will be responsible for analytical changes to the FE models. The FE dynamic models, generated under Army funding during the develop-mental phase of the ACAP program, were further improved under funding of the NASA DAMVIBS program. However, the thrust of shake testing performed during the developmental phase was oriented towards the usefulness of the models to 15 Hz and below. In the correlation to be performed, the test vehicles will be stripped down to the basic structure. The inertia of the components removed will be substituted with concentrated masses. Upon successful correlation of the basic configuration, components will be installed and correlation efforts repeated. The goal is to achieve satisfactory correlation at modal and force response frequencies up to 40 Hz.

TITLE: Composite Airframe Design for Weapons Interface

RESPONSIBLE INDIVIDUAL: E. Austin
U.S. Army ARTA (AVSCOM)
Aviation Applied Technology Directorate
SAVRT-TY-ATS
Ft. Eustis, VA 23604-5577
(804) 878-3822

PRINCIPAL INVESTIGATORS: J. Moffatt
U.S. Army ARTA (AVSCOM)
Aviation Applied Technology Directorate
SAVRT-TY-ATS
Ft. Eustis, VA 23604-5577
(804) 878-2377

OBJECTIVE: The effect of 20-30mm weapon firing in close proximity to composite airframe is investigated. Effects of weapon-induced pressure and thermal environments on weight tradeoffs for structural design are investigated.

U. S. ARMY AVIATIONS SYSTEMS COMMAND
US ARMY RESEARCH & TECHNOLOGY ACTIVITY
NASA-LANGLEY RESEARCH CENTER

TITLE: Basic Research in Structures

RESPONSIBLE INDIVIDUAL: Dr. F. D. Bartlett, Jr.
U.S. Army ARTA
Aerostructures Directorate
Mail Stop 266
NASA Langley Research Center
Hampton, VA 23665-5225
(804) 864-3960

PRINCIPAL INVESTIGATORS: Dr. T.K. O'Brien, Dr. R.L. Boitnott, G.L. Farley, M.W. Nixon

OBJECTIVE: The objectives and scope of this research are to investigate and explore structures technologies which exploit advanced materials for improved structural performance, develop superior analyses for composites design, and devise automated processes for inspecting and manufacturing rotorcraft structures. This is accomplished, in conjunction with NASA Langley, by conducting basic research of composite and metallic materials to understand and improve fatigue resistance, fracture toughness, crash-worthiness, and internal noise transmission as well as to develop more efficient and damage tolerant structural forms for rotorcraft applica-tions.

TITLE: Structures Technology Applications

RESPONSIBLE INDIVIDUAL: Dr. F.D. Bartlett, Jr.
U.S. Army ARTA
Aerostructures Directorate
Mail Stop 266
(804) 865-2866

PRINCIPAL INVESTIGATORS: Dr. R.L. Boitnott, M.W. Nixon, G.L. Farley, D.J. Baker

OBJECTIVE: The goals of this research are to explore and demonstrate innovative structural concepts and design methodologies which will provide lighter, safer, and more survivable structures for rotorcraft. This is achieved through jointly-sponsored Army/NASA investigations which establish improved structural integrity and crashworthiness, validate superior analytical capabilities, and demonstrate lower cost manufacturing processes. The emphasis of this research is to provide proven technology to the rotorcraft industry and the U.S. Army for applications to future air vehicle systems.

OFFICE OF NAVAL RESEARCH
MECHANICS DIVISION
ARLINGTON VA 22217-5000

GRANTS AND CONTRACTS

FAILURE OF THICK COMPOSITE LAMINATES
N00014-88-F-0044
February 88 - January 93

Scientific Officer: Dr. Yapa D.S. Rajapakse
Office of Naval Research
Mechanics Division, Code 1132SM
Arlington, VA 22217-5000
(202) 696-4405, Autovon 226-4405

Principal Investigator: Dr. R.M. Christensen
Lawrence Livermore National Laboratory
PO Box 808
Livermore, CA 94550
(415) 422-7236

Objective: Research will be conducted into the mechanics of failure of composite materials, with emphasis on physically-based failure criteria for thick composites laminates.

NONDESTRUCTIVE EVALUATION AND DAMAGE ACCUMULATION OF COMPOSITES
N00014-90-J-1724
April 87 - September 92

Scientific Officer: Dr. Yapa D.S. Rajapakse
Office of Naval Research
Mechanics Division, Code 1132SM
Arlington, VA 22217-5000
(202) 696-4405, Autovon 226-4405

Principal Investigator: Prof. I. M. Daniel
Northwestern University
Department of Civil Engineering
Evanston, IL 60201
(312) 491-5649

Objective: Research will be conducted to understand the process of damage growth in thick composites laminates subjected to complex loading states and fatigue. Nondestructive methods for damage characterization will be developed.

ENVIRONMENTAL EFFECTS AND ENVIRONMENTAL DAMAGE IN THERMOPLASTIC COMPOSITES
N00014-90-J-1556
Jan 90 - Dec 92

Scientific Officer: Dr. Yapa D.S. Rajapakse
Office of Naval Research
Mechanics Division, Code 1132SM
Arlington, VA 22217-5000
(202) 696-4405, Autovon 226-4405

Principal Investigator: Prof. Y. Weitsman
University of Tennessee
Dept. of Eng. Sci. & Mechanics
Knoxville, TN 37996-2030
(615) 974-5460

Objective: Research will be conducted into the effects of constant and cyclic pressure on moisture absorption and moisture-induced damage in thermoplastic composites. The development of residual stresses during processing will also be investigated.

DYNAMIC MATRIX CRACKING AND DELAMINATION IN COMPOSITE LAMINATES SUBJECTED TO IMPACT LOADING
N00014-90-J-1666
July 84 - November 92

Scientific Officer: Dr. Yapa D.S. Rajapakse
Office of Naval Research
Mechanics Division, Code 1132SM
Arlington, VA 22217-5000
(202) 696-4405, Autovon 226-4405

Principal Investigator: Prof. C.T. Sun
Purdue University
School of Aeronautics and Astronautics
West Lafayette, IN 47907
(317) 494-5130

Objective: The propagation of damage in composite laminates due to impact loading conditions will be investigated using theoretical and experimental techniques. Dynamic delamination models will be established. Comcepts for controlling impact damage will be explored, including the use of soft adhesive strips.

THERMOMECHANICAL BEHAVIOR OF HIGH TEMPERATURE COMPOSITES
N00014-89-J-3107
March 85 - July 91

Scientific Officer: Dr. Yapa D.S. Rajapakse
Office of Naval Research

Mechanics Division, Code 1132SM
Arlington, VA 22217-5000
(202) 696-4405, Autovon 226-4405

Principal Investigator: Prof. G.J. Dvorak
Rensselaer Polytechnic Institute
Department of Civil Engineering
Troy, NY 12181
(518) 276-6943

Objective: Investigations of the thermomechanical response, damage growth and fracture in metal matrix composites and intermetallic matrix composites will be conducted using analytical and experimental techniques. Local stress states caused during fabrication and by thermal changes in service, in elastic time-dependent behavior, and static and fatigue damage will be explored.

QUANTITATIVE ULTRASONICS MEASUREMENTS IN COMPOSITES
N00014-90-J-1273
July 85 - September 92

Scientific Officer: Dr. Yapa D.S. Rajapakse
Office of Naval Research
Mechanics Division, Code 1132SM
Arlington, VA 22217-5000

Principal Investigator: Prof. W. Sachse
Cornell University
Dept. of Theoretical and Applied Mechanics
Ithaca, NY 14853
(609) 255-5065

Objective: Research will be conducted to establish quantitative active and passive ultrasonic measurement techniques for characterizing the microstructure and mechanical properties as well as the dynamics of deformation processes in composite materials.

DYNAMIC BEHAVIOR OF FIBER AND PARTICLE REINFORCED COMPOSITES
N00014-86-K-0280
October 86 - December 93

Scientific Officer: Yapa D.S. Rajapakse
Office of Naval Research
Mechanics Division, Code 1132SM
Arlington, VA 22217-5000
(202) 696-4405, Autovon 226-4405

Principal Investigator: Prof. S. K. Datta
University of Colorado
Department of Mechanical Engineering
Boulder, CO 80309
(303) 492-1139

Objective: Research will be conducted into the diffraction of elastic waves by cracks and other inhomogeneities in laminated fiber reinforced composites. Investigations of dynamic material properties of fiber and particle reinforced metal-matrix composites will be conducted, accounting for interfacial effects.

IMPACT RESPONSE AND QNDE OF LAYERED COMPOSITES
N00014-87-K-0351
April 87 - December 91

Scientific Officer: Dr. Yapa D.S. Rajapakse
Office of Naval Research
Mechanics Division, Code 1132SM
Arlington, VA 22217-5000
(202) 696-4405, Autovon 226-4405

Principal Investigator: Prof. A.K. Mal
University of California, Los Angeles
Dept. of Mechanical, Aerospace & Nuclear Engineering
Los Angeles, CA 90024
(213) 825-5481

Objective: Research will be conducted into wave propagation in composite laminates, with the focus on dynamic loading conditions and theoretical aspects of quantitative acoustic microscopy. The Leaky Lamb Wave technique will be utilized for the characterization of elastic properties and defects in composites. The use of ultrasonic techniques for interfaces and interfacial regions will be explored.

MICROMECHANICS OF COMPOSITES
N00014-90-J-1377
October 90 - September 91

Scientific Officer: Dr. Yapa D.S. Rajapakse
Office of Naval Research
Mechanics Division, Code 1132SM
Arlington, VA 22217-5000
(202) 696-4405, Autovon 226-4405

Principal Investigator: Prof. B. Budiansky
Harvard University
Division of Applied Science
Cambridge, MA 02138
(617) 495-2849

Objective: Research will be conducted into the micromechanical enhancement of the fracture toughness of ceramics and intermetallics by the incorporation of toughening agents such as fibers, whiskers, ductile particles and phase-transforming

particles. Models will be established for the compression failure of polymer matrix composites.

MECHANICS OF INTERFACE CRACKS AND COMPOSITES
N00014-90-J-1380
November 87 - November 91

Scientific Officer: Dr. Yapa D.S. Rajapakse
Office of Naval Research
Mechanics Division, Code 1132SM
Arlington, VA 22217-5000
(202) 696-4405, Autovon 226-4405

Principal Investigator: Prof. C.F. Shih
Brown University
Division of Engineering
Providence, RI 02912
(401) 863-2868

Objective: Research will be conducted to provide a fundamental understanding of the behavior of interface cracks in bimaterial elastic-plastic systems. The stress and strain fields around such cracks will be studied at both the continuum and polycrystalline slip theory levels.

FRACTURE MECHANICS OF INTERFACIAL ZONES IN BONDED MATERIALS
N00014-89-J-3188
Septmber 89 - August 91

Scientific Officer: Dr. Yapa D.S. Rajapakse
Office of Naval Research
Mechanics Division, Code 1132SM
Arlington, VA 22217-5000
(202) 696-4405, Autovon 226-4405

Principal Investigator: Prof. F. Erdogan
Lehigh University
Dept. of Mechanical Engr. & Mechanics
Bethlehem, PA 18015
(215) 758-3020

Objective: Research will be conducted into the micromechanics aspects of failure of composites, accounting for realistic interfacial zones. Models will be established for crack propagation in interfacial regions with continuously varying mechanical properties.

SYNCHROTRON X-RAY MICROTOMOGRAPHY FOR COMPOSITES
N00014-89-C-0076
April 89 - March 90

Scientific Officer: Dr. Yapa D.S. Rajapakse
Office of Naval Research
Mechanics Division, Code 1132SM
Arlington, VA 22217-5000
(202) 696-4405, Autovn 226-4405

Principal Investigator: Dr. A.S. Krieger
Radiation Science, Inc
PO Box 293
Belmont, MA 02173
(617) 494-0335

Objective: The nondestructive technique of three dimensional synchrotron x-ray microtomography will be used for the assessment of internal structure and defects in composite materials and composite interfaces.

NATIONAL CENTER FOR COMPOSITE MATERIALS RESEARCH
p400013f101
September 86 - September 91

Scientific Officer: Dr. R.J. Jones
Office of Naval Research
Mechanics Division, Code 1132SM
Arlington, VA 22217-5000
(202) 696-4305, Autovon 226-4305

Principal Investigator: Prof. J. Economy
University of Illinois
National Center for Composite Materials Research
Urbana, IL 61801
(215) 333-1835

Objective: Under ONR-URI sponsorship, a National Center for Composite Materials Research was established to conduct a well structured, multidisciplinary research program in composites spanning the disciplines of solid mechanics, materials science, chemistry and surface physics. Initial emphasis will be on critical research issues associated with the use of thick composites for ship and submarine structures.

FAILURE MECHANICS OF THICK COMPOSITES
N00014-85-K-0439
October 85 - September 93

Scientific Officer: Dr. Yapa D.S. Rajapakse
Office of Naval Research
Mechanics Division, Code 1132SM
Arlington, VA 22217-5000
(202) 696-4405, Autovn 226-4405

Principal Investigator: Prof. S.N. Atluri
Georgia Institute of Technology
Dept. of Civil Engineering
Atlanta, GA 30332=3055
(404) 894-2758

Objective: Research will be conducted into three-dimensional aspect of damage and failure in composites. Compression failure will be investigated.

OPTICAL MAPPING OF DEFORMATION FIELDS AROUND INTERFACE CRACKS
N00014-82-K-0566
January 87 - December 90

Scientific Officer: Dr. Yapa D.S. Rajapakse
Office of Naval Research
Mechanics Division, Code 1132SM
Arlington, VA 22217-5000
(202) 696-4405, Autovn 226-4405

Principal Investigator: Prof. F.P. Chiang
State Univesity of New York
Dept. of Mechanical Engineering
Stony Brook, NY 11794-2300
(516) 246-6768

Objective: The optical techniques of moire interferometry and laser speckle interferometry will be used to determine 2D and 3D deformations in the vicinity of interface cracks.

NAVAL RESEARCH LABORATORY
WASHINGTON, DC 20375-5000

IN-HOUSE

SIMULATION OF STRUCTURAL RESPONSE OF DAMAGED COMPOSITE SHIP COMPONENTS
October 86 - September 92

Principal Invesigator: **Dr. Phillip Mast**
Naval Research Laboratory
Code 6383
Washington, DC 20375-5000
(202) 767-2165, Autovon 297-2165

Objective: Develop and apply an advanced simulation capability for predicting the effect of damage on the structural response of naval components made with fiber reinforced composites.

NAVAL RESEARCH LABORATORY
WASHINGTON, DC 20375-5000

CONTRACTS

DYNAMIC BEHAVIOR OF COMPOSITES
N00014-86-C-2580
October 86 - March 91

Scientific Officer: **Mr. Irvin Wolock**
Naval Research Laboratory
Washington, DC 20375-5000
(202) 767-2567, Autovon 297-3567

Principal Investigator: **Dr. Longin B. Greszczuk**
McDonnell Douglas Astronautics Company
5301 Bolsa Avenue
Huntington Beach, CA 92647
(714) 896-3810

Objective: Develop a capability to predict the effects of large area dynamic loading, such as that due to an underwater explosion, on the mechanical response of composite materials and structures.

NAVAL AIR DEVELOPMENT CENTER
WARMINSTER, PA 19874-5000

IN-HOUSE

STRUCTURAL DAMPING
October 87 - December 90

Project Engineer: Dr. D.J. Barrett
Naval Air Development Center
AVCSTD/6043
Warminster, PA 18974-5000
(215) 441-1130, Autovon 441-1330

Objective: Improve the damping properties of structures through the redesign of basic structural components as composites of stiffness and damping materials.

ANALYTICAL MODELING OF COMPOSITE INTERFACE MECHANICS
April 88 - September 91

Project Engineer: Dr. H.C. Tsai
Naval Air Development Center
AVCSTD/6043
Warminster, PA 18974-5000
(215) 441-1287, Autovon 441-1287

Objective: Understand how interface failure mechanisms develop and influence the properties of resin matrix composites and devise non-linear micromechanics models to describe the behavior at the interface region.

METAL MATRIX CRACK INITIATION/PROPAGATION
September 85 - December 90

Project Engineer: Dr. H.C. Tsai
Naval Air Development Center
AVCSTD/6043
Warminster, PA 18974-5000
(215) 441-1287, Autovon 441-1287

Objective: Characterize the crack initation/propagation mechanics of silicon carbide/titanium metal matrix composites as applied to landing gear and arrestor hooks in the naval shipboard environment.

DAVID TAYLOR RESEARCH CENTER
BETHESDA, MD 20084-5000
ANNAPOLIS, MD 21842

IN-HOUSE

COMPRESSION RESPONSE OF THICK-SECTION COMPOSITE MATERIALS
October 86 - September 91

Principal Investigator: E.T. Camponeschi, Jr.
David Taylor Research Center, Code 2844
Annapolis, MD 21842
(301) 267-2165, Autovon 281-2165

Objective: Develop an understanding of compression failure for thick section composites.

BEHAVIOR OF COMPOSITES SUBJECTED TO UNDERWATER EXPLOSIVE LOADING
January 87 - September 91
Principal Investigator: Erik Rasmussen
David Taylor Research Center, Code 1720
Bethesda, MD 20084-5000
(301) 227-1656, Autovon 287-1656

Objective: Develop the analytical and experimental techniques required to assess the dynamic capabilities of proposed composite submarine pressure hull structural and material concepts.

COMPOSITE PRESSURE HULL PENERATION AND JOINT DESIGN
June 88 - September 91

Principal Investigator: M. Brown
David Taylor Research Center, Code 1720.2
Bethesda, MD 20084-5000
(310) 227-1706, Autovon 287-1706

Objective: Develop structurally efficient joint, peneration, and reinforcement concepts for composite pressure hulls; the analytical capability to predict the structural response of these concepts; the experimental capability to verify the validity of the analytical procedures.

COMPOSITE STRUCTURES FOR SURFACE SHIPS
October 85 - September 93

Principal Investigator: M. Critchfield
David Taylor Research Center
Bethesda, MD 20084-5000
(301) 227-1769, Autovon 287-1769

Objective: Develop the basic technology to support the applications of composites to naval ship structures including design and analytic methods in structural joint and attachments, and to demonstrate the feasibility of using FRP composites for surface ship structural applications such as deckhouses, stacks and masts, and secondary structures.

WRIGHT RESEARCH AND DEVELOPMENT CENTER
MATERIALS LABORATORY

IN-HOUSE

ADVANCED COMPOSITES
WORK UNIT DIRECTIVE (WUD) NUMBER 45
89 October - 91 October

WUD Leader: James M. Whitney
Materials Laboratory
Wright Research and Development Center
WRDC/MLBM
Wright-Patterson AFB OH 45433-6533
(513) 255-9097, AUTOVON: 785-9097

Objective: The objective of the long term thrust is to develop understanding of deformation and failure process of composite laminates. The short term objectives include the following: (a) understanding failure mechanisms under compression loading; (b) failure of brittle matrix composites.

CONTRACTS

IMPROVED TECHNOLOGY FOR ADVANCED COMPOSITE MATERIALS
F33615-87-C-5239
15 Sep 87 - 1 Feb 92

Project Engineer: Ken Johnson
Materials Laboratory
Wright Research and Development Center
WRDC/MLBC
Wright-Patterson AFB OH 45433-6533
(513) 255-6981, AUTOVON: 785-6981

Principal Investigator: Rebecca C. Schiavone
University of Dayton Research Institute
300 College Park Avenue
Dayton OH 45469

Objective: The objective of this program is to investigate from both an experimental and an analytical standpoint the potential of new and/or modifications of existing matrix materials and reinforcements/product forms for use in advanced composite materials, including processing/mechanical property relationships. Such materials are subsequent candidates for use in advanced aircraft and aerospace structural applications.

MICROMECHANICS OF COMPOSITE FAILURE
F33615-88-C-5420
1 Oct 88 - 30 Sep 92

Project Engineer: Nicholas J. Pagano
Materials Laboratory
Wright Research and Development Center
WRDC/MLBM
Wright-Patterson AFB OH 45433-6533
(513) 255-6762, AUTOVON 785-6762

Principal Investigator: Som R. Soni
AdTech Systems Research Inc
1342 N. Fairfield Road
Dayton OH 45432

Objective: The objective of this program is to provide exploratory development in thermomechanical response, model material system development composite processing, and failure mechanisms investigations of composite and related constituent materials.

DEVELOPMENT OF ULTRA-LIGHTWEIGHT MATERIALS-N
F33615-88-C-5447
29 Apr 88 - 1 Jul 91

Project Engineer: Lt Suzanne Guihard
Materials Laboratory
Wright Research and Development Center
WRDC/MLBC
Wright-Patterson AFB OH 45433-6533
(513) 255-9728, AUTOVON: 785-9728

Principal Investigator: Anne R. Beck
Northrop Corporation
Aircraft Division
One Northrop Avenue
Hawthorne CA 90250

Objective: To demonstrate the potential for advanced ultra-lightweight (ULW) materials and associated processes that will permit a fifty percent reduction in the structural weight of state-of-the-art (SOTA) high-performance aircraft that currently utilize up to ten percent of advanced composite materials in their structures.

DEVELOPMENT OF ULTRA-LIGHTWEIGHT MATERIALS-M
F33615-88-C-5452
13 May 88-15 Jul 91

Project Engineer: Lt Suzanne Guihard
Materials Laboratory
Wright Research and Development Center
WRDC/MLBC
Wright-Patterson AFB OH 45433-6533
(513) 255-9728, AUTOVON: 785-9728

Principal Investigator: Gail L. Dolan
McDonnell Douglas Corporation
McDonnell Douglas Company
PO Box 516
St Louis MO 63166

Objective: To demonstrate the potential for advanced ultra-lightweight (ULW) materials and associated processes that will permit a fifty percent reduction in the structural weight of state-of-the-art (SOTA) high-performance aircraft that currently utilize up to ten percent of advanced composite materials in their structures.